TRANSACTIONS
American Society for Metals
VOL. XXXVI 1946

TECHNICAL PROGRAM AND REPORTS OF OFFICERS AMERICAN SOCIETY FOR METALS—27TH ANNUAL CONVENTION, CLEVELAND, FEBRUARY 4 to 8, 1946

FOR purposes of record and for the benefit of members who were not in attendance at the Twenty-seventh Annual Convention of the Society, held in Cleveland, February 4 to 8, 1946, the Programs of the Technical Papers, the Round Table Discussions, and Educational Lectures together with the Reports of Officers for 1945 are herewith published in full.

TECHNICAL PAPERS PROGRAM

MONDAY, FEBRUARY 4

Lattice Room, Hotel Statler—9:30 A. M.

Session on Aluminum Alloys

Joint Chairmen—L. W. Davis, Aluminum Company of America, and H. Y. Hunsicker, Aluminum Company of America

Drawability of Aluminum Alloys at Elevated Temperatures. Part I—Deep Drawing Cylindrical Cups, by D. M. Finch, S. P. Wilson and J. E. Dorn, University of California.

Deep Drawing Aluminum Alloys at Elevated Temperatures. Part II—Deep Drawing Boxes, by D. M. Finch, S. P. Wilson and J. E. Dorn, University of California.

New Aluminum Alloys Containing Small Amounts of Beryllium, by R. H. Harrington, General Electric Co.

Grand Ballroom, Hotel Statler—9:30 A. M.

Session on Heat Treatment of Alloy Steels

Joint Chairmen—L. S. Bergen, Bergen Precision Castings Co., and G. M. Cover, Case School of Applied Science

The Partition of Molybdenum in Hypoeutectoid Iron-Carbon-Molybdenum Alloys, by Fred E. Bowman, Climax Molybdenum Co.

The Effect of Variations in Composition and Heat Treatment on Some Properties of 4 to 6% Chromium Steel Containing Molybdenum and Titanium, by Geo. F. Comstock, Titanium Alloy Mfg. Co.

Iron-Manganese Alloys—the Properties of Cold-Worked and Heat Treated Alloys Containing 1 to 7% Manganese, by R. S. Dean, J. R. Long, T. R. Graham and R. G. Feustel, Bureau of Mines.

Euclid Ballroom, Hotel Statler—9:30 A. M.

Session on Carburizing and Decarburizing

Joint Chairmen—H. W. McQuaid, Republic Steel Corp., and F. N. Rhines, Carnegie Institute of Technology

The Application of Ms Points to Case Depth Measurement, by E. S. Rowland and S. R. Lyle, Timken Roller Bearing Co.

A Mechanism of the Surface Decarburization of Steel, by W. A. Pennington, Carrier Corp.

Graphite in Cold-Rolled Subcritically Annealed Hypoeutectoid Steels, by M. A. Hughes and J. G. Cutton, Carnegie-Illinois Steel Corp.

TUESDAY, FEBRUARY 5

Euclid Ballroom, Hotel Statler—9:30 A. M.

Session on Research

Joint Chairmen—O. E. Harder, Battelle Memorial Institute, and M. F. Judkins, Firth-Sterling Steel Co.

High Forging Temperatures Revealed by Facets in Fracture Tests, by Robert Strohm, Copperweld Steel Co., and W. E. Jominy, Chrysler Corp.

Critical Points of SAE 4340 Steel as Determined by the Dilatometric Method, by D. Niconoff, Republic Steel Corp.

Gas Evolution from Cast Steel at Room Temperature, by H. H. Johnson, L. H. Arner and H. A. Schwartz, National Malleable & Steel Castings Co.

Grand Ballroom, Hotel Statler—9:30 A. M.

Session on Stresses in Metals

Joint Chairmen—A. E. Focke, Diamond Chain & Manufacturing Co., and M. Gensamer, Pennsylvania State College

The Effects of Combined Stresses and Low Temperatures on the Mechanical Properties of Some Nonferrous Metals, by D. J. McAdam, Jr., G. W. Geil and R. W. Mebs, National Bureau of Standards.

Temper Brittleness, by John H. Hollomon, Watertown Arsenal.

Fracture of Metals Under Combined Stresses, by D. J. McAdam, Jr., National Bureau of Standards.

Lattice Room, Hotel Statler—9:30 A. M.

Session on Nonferrous Alloys

Joint Chairmen—W. E. Mahin, Armour Research Institute, and P. T. Stroup, Aluminum Company of America

Soft Soldering, by M. E. Fine and R. L. Dowdell, University of Minnesota.

The Cold Working and Heat Treatment of a 10-Carat Gold Alloy, by Vernon H. Patterson and B. N. Iannone, Bausch & Lomb Optical Co.

Tellurium Corewashes, by James O. Vadeboncoeur, General Motors Corp.

TUESDAY, FEBRUARY 5

Ballroom, Public Auditorium—2:00 P. M.

Session on Chromium and Stainless Steels

Joint Chairmen—E. G. Mahin, University of Notre Dame, and A. O. Schaefer, The Midvale Co.

Investigation of a Type of Failure of 18-8 Stabilized Stainless Steel, by W. C. Kahn, New York Testing Laboratories, H. Oster, Republic Aviation Corp., and R. Wachtell, Apparatus Research and Development Laboratory.

The Influence of Carbon Content Upon the Transformations in 3% Chromium Steel, by Taylor Lyman, American Society for Metals, and A. R. Troiano, University of Notre Dame.

Effect of Nickel on Physical Properties and Thermal Characteristics of Some Cast Chromium-Molybdenum Steels, by N. A. Ziegler and W. L. Meinhart, Crane Co.

Clubroom B, Public Auditorium—2:00 P. M.

Session on Quenching of Steels

Joint Chairmen—D. T. Doll, Case School of Applied Science, and W. E. Jominy, Chrysler Corp.

Factors Affecting the Hardenability of Boron-Treated Steels, by R. A. Grange, U. S. Steel Corp. Research Laboratories, and T. M. Garvey, Carnegie-Illinois Steel Corp.

Quenching of Steel Balls and Rings, by Victor Paschkis, Columbia University.

Mass Temperature Effects on Quenching 36% Cobalt Magnet Steel, by Benjamin Falk, Simonds Saw & Steel Co.

WEDNESDAY, FEBRUARY 6

Grand Ballroom, Hotel Statler—9:30 A. M.

Reports of Officers of American Society for Metals for 1945

Edward de Mille Campbell Memorial Lecture, by M. Gensamer, Pennsylvania State College.

Vsevolod N. Krivobok, Chairman

WEDNESDAY, FEBRUARY 6

Clubroom B, Public Auditorium—2:00 P. M.

Session on Inspection

Joint Chairmen—A. A. Bates, Westinghouse Electric Corp., and H. B. Pulsifer, American Metal Treating Co.

Anti-Reflection Films for Metallographic Objectives, by James R. Benford, Bausch & Lomb Optical Co.

Detection, Causes and Prevention of Injury in Ground Surfaces, by L. P. Tarasov, Norton Co.

The Practical Application of Statistical Methods in a Quality Control Program, by W. T. Rogers, National Tube Co.

Ballroom, Public Auditorium—2:00 P. M.

Session on Induction Hardening

Joint Chairmen—W. E. Benninghoff, Ohio Crankshaft Co., and W. G. Johnson, Caterpillar Tractor Co.

Stress Comparisons by Correlation With High Frequency Magnetic and Eddy Current Losses, by P. E. Cavanagh, Allen B. DuMont Laboratories.

Metallurgical Characteristics of Induction-Hardened Steel, by James W. Poynter, Army Air Forces, Wright Field.

Induction Hardening and Austenitizing Characteristics of Several Medium Carbon Steels, by D. L. Martin, General Electric Co., and W. G. Van Note, North Carolina State College.

ROUND TABLE DISCUSSIONS

MONDAY, FEBRUARY 4

Music Hall, Public Auditorium—8:30 P. M.

Decomposition of Austenite

Chairman—Capt. J. H. Hollomon, Ordnance Department, Watertown Arsenal.

Panel Members—Robert F. Mehl, head, department of metallurgy, Carnegie Institute of Technology.

Clarence Zener, Institute for the Study of Metals, University of Chicago.

W. H. Brandt, assistant manager, materials engineering department, Westinghouse Electric Corp.

E. S. Davenport, assistant to vice-president, United States Steel Corp.

Morris Cohen, department of metallurgy, Massachusetts Institute of Technology.

TUESDAY, FEBRUARY 5

Music Hall, Public Auditorium—8:30 P. M.

Super-Alloys for High Temperature Service in Gas Turbines and Jet Engines

Chairman—C. T. Evans, Jr., chief metallurgist, Elliott Co.

Panel Members—Howard Cross, Battelle Memorial Institute.

Russell Franks, chief metallurgist, Union Carbide and Carbon Research Laboratories.

N. L. Mochel, metallurgical engineer, Westinghouse Electric Corp.

Gunther Mohling, Allegheny Ludlum Steel Corp.

W. L. Badger, metallurgical section, Thompson Laboratory, General Electric Co.

F. S. Badger, metallurgist, Haynes Stellite Co.

WEDNESDAY, FEBRUARY 6

Music Hall, Public Auditorium—8:30 P. M.

Atomic Energy and Its Implications (Nucleonics)

Chairman—Cyril Stanley Smith, director of Metals Institute, University of Chicago.

Panel Members—S. K. Allison, Institute of Nuclear Studies, University of Chicago.

Zay Jeffries, vice-president, General Electric Co.

A. B. Kinzel, vice-president, Union Carbide and Carbon Research Laboratories.

John Chipman, professor of metallurgy, Massachusetts Institute of Technology.

EDUCATIONAL LECTURES

Held in Clubrooms A, B, and C and the Ballroom, Public Auditorium

THURSDAY, FEBRUARY 7—9:00 A.M.

Lecture 1 on Magnesium: *Extractive Metallurgy,* by L. M. Pidgeon, University of Toronto.

Lecture 1 on Induction Heating: *Principles and Theory of High Frequency Heating,* by H. B. Osborn, Jr., Ohio Crankshaft Co.

Lecture 1 on Surface Stressing: *The Problem Defined,* by H. F. Moore, University of Illinois.

Lecture 1 on Corrosion: *Basic Principles of Metallic Corrosion,* by Carl W. Borgmann, University of Colorado.

THURSDAY, FEBRUARY 7—10:30 A.M.

Lecture 2 on Magnesium: *Magnesium Structural Design,* by J. C. Mathes, Dow Chemical Co.

Lecture 2 on Induction Heating: *Induction Heating Circuits and Frequency Generation,* by P. H. Brace, Westinghouse Electric Corp..

Lecture 2 on Surface Stressing: *Measurement of Surface Stresses,* by W. M. Murray, Society for Experimental Stress Analysis.

Lecture 2 on Corrosion: *Effect of Composition and Environment on Corrosion of Iron and Steel,* by C. P. Larrabee, Carnegie-Illinois Steel Corp.

THURSDAY, FEBRUARY 7—4:00 P.M.

Lecture 3 on Magnesium: *Castings,* by N. E. Woldman, Bendix Aviation Corp.

Lecture 3 on Induction Heating: *Practical Applications of the Motor-Generator Type of Induction Heating (Frequencies up to 10,000 Cycles),* by W. G. Johnson, Caterpillar Tractor Co.

Lecture 3 on Surface Stressing: *Methods of Applying and Tests Used Including Carburizing and Nitriding,* by J. O. Almen, General Motors Corp.

Lecture 3 on Corrosion: *Corrosion Resistance of Stainless Steels and High Nickel Alloys,* by W. O. Binder, Union Carbide & Carbon Research Laboratories.

FRIDAY, FEBRUARY 8—9:00 A.M.

Lecture 4 on Magnesium: *A Survey of Wrought Magnesium Alloy Fabrication,* by J. V. Winkler, Dow Chemical Co. Presented by C. E. Lehnhardt, Dow Chemical Co.

Lecture 4 on Induction Heating: *Practical Applications of High Frequency Induction Heating (100,000 Cycles and Up),* by J. W. Cable, Induction Heating Corp.

Lecture 4 on Surface Stressing: *Stressing Axles and Other Railroad Equipment by Cold Rolling,* by O. J. Horger, Timken Roller Bearing Co.

Lecture 4 on Corrosion: *Corrosion of Light Metals (Aluminum and Magnesium),* by E. H. Dix, Jr., Aluminum Co. of America.

FRIDAY, FEBRUARY 8—10:30 A.M.

Lecture 5 on Magnesium: *Corrosion and Protection of Magnesium,* by W. S. Loose, Dow Chemical Co.

Lecture 5 on Induction Heating: *A Comparison of Induction Heating with Other Methods of Heat Treating,* by T. E. Eagan, Cooper-Bessemer Corp.

Lecture 5 on Surface Stressing: *Progressive Stress Damage,* by Peter R. Kosting, Watertown Arsenal.

Lecture 5 on Corrosion: *Copper and Copper Alloys in Corrosive Environments,* by H. L. Burghoff, Chase Brass & Copper Co.

REPORTS OF OFFICERS OF AMERICAN SOCIETY FOR METALS FOR 1945

At 9:30 A.M. Wednesday, February 6, 1946, members of the Society convened in the Grand Ballroom of the Hotel Statler to hear the reports of officers of the Society for the year 1945. Although the annual meeting of the Society had been held in conjunction with the November 2, 1945, meeting of the Cleveland Chapter, the reading of these reports was deferred until the holding of the 27th annual convention. The election and installation of officers for 1946, however, was consummated at that meeting in accordance with the provisions of the constitution. The report of the election of officers follows the reports of officers in this volume of TRANSACTIONS.

At the conclusion of the reading of the reports of officers, the meeting was adjourned but was immediately reconvened for presen-

tation of the Edward deMille Campbell Memorial Lecture. This year the Society was honored by having Dr. Maxwell Gensamer, Professor of Metallurgy and Head, Department of Mineral Technology, School of Mineral Industries, Pennsylvania State College, State College, Pa., present his lecture entitled *"Strength and Ductility"*.

The lecture is printed in full in this volume of TRANSACTIONS.

ANNUAL PRESIDENT'S ADDRESS

KENT R. VAN HORN, *President*

Twenty-seventh Annual Convention, Cleveland, February 4 to 8, 1946

WITH the end of the war and the start of reconversion, it seems appropriate to take stock of your Society's position in the metal industry, and to review briefly the accomplishments of the American Society for Metals in World War II and to consider its future opportunities for service.

Certainly, any society worth its salt in peacetime is worth its weight in gold in time of war—and I believe that the American Society for Metals proved its worth during the last four years. In fact, your Society started the transition from peace to wartime service in September, 1939.

It was then that Army and Navy authorities asked your Society to assist in the educational training of inspectors at the Philadelphia Arsenal, Navy Yard and Aircraft Factory. As a result, a series of ten lectures on inspection of metals was arranged and presented to an attendance of 800 prospective inspectors. Thereafter, these training lectures were presented in many other cities and were attended by some 10,000 enrollees. Your Board of Trustees also authorized the preparation and publication of an inspection textbook to be sold at cost. This book—and I know many of you are familiar with it—was *Inspection of Metals*. It was printed four times and was distributed to more than 10,000 men.

In June, 1941, the Office of Production Management—later the War Production Board—asked your Society to appoint committees and prepare recommended practices for the preparation, use and treatment of molybdenum high speed steels. As you know, the shortage of tungsten required a reduction in the use of this element

and an increase in the use of molybdenum for high speed steels. Within two months, your Society completed this work and submitted its report. Thousands of copies were prepared and distributed to industry, to the technical and trade press of America, Canada and England. In addition, many of the chapters arranged special meetings to assist materially in the rapid conversion from tungsten to molybdenum tool steels.

When war came in December, 1941, your Society formed War Products Advisory Committees—forty-one chapters participating in this concerted effort to give counsel and advice without cost or obligation to all manufacturing plants with war problems involving metals. Every branch of metallurgical engineering was represented by American Society for Metals experts on each of these committees. Many of you contributed valuable time. Meeting at regular intervals, manufacturers were invited to bring their problems before these boards for entirely free advice. Mailing lists were compiled and letters, leaflets and other literature were sent to all plants in the various areas. The measure of war aid given by these committees may never be determined, but it is a fact that scores of problems were presented and solved by the after-hours work of A.S.M. members. You may know, too, that when the National Emergency Steels were developed to conserve scarce alloys, twenty-eight of the committee of twenty-nine men who conceived these steels were members of your Society. Once available, conferences were held in Washington to determine plans by which the War Products Advisory Committees could distribute information on these new steels and secure consumer acceptance of them. The speed and energy with which these committees acted received much praise. As a further step in expediting NE steels, your Society issued a 40-page booklet compiling information and data on their use, together with a letter of commendation from the WPB Chairman, Donald Nelson. This booklet had five printings, two revisions and over 12,000 copies were distributed.

In addition to all these special activities, the regular services of the American Society for Metals were geared to wartime requirements. Special round table sessions at the Metal Congress were devoted to important practical problems of war, such as metal conservation and substitution, employee training, tool repair, etc. Manufacturers who exhibited in the Metals Expositions brought engineers and metal experts on a free consulting basis to answer the production problems of the industry.

Your monthly magazine, METAL PROGRESS, devoted itself exclusively to developments that would aid in the war effort. Many articles were presented on metal shortages, on strategic and critical metals, and on substitutes. An editorial feature, "War Products Consultation", was inaugurated to deal with a specific problem or inquiry and to present a solution by a noted expert. Your other publications—the METALS HANDBOOK, the METALS REVIEW and TRANSACTIONS of the American Society for Metals—each was helpful in meeting the demands of wartime production.

These efforts of your Society were recognized by the Ordnance Department of the United States Army in July, 1944, by the Distinguished Service Award. Major General G. M. Barnes, Chief of Research and Development, presented the Award at the Annual Banquet in Cleveland for "outstanding and meritorious engineering advisory services, in war and peace, for the development, manufacture and maintenance of Ordnance matériel".

In short, the American Society for Metals adjusted itself to a War of Metals. It enlisted for the duration, and it has emerged from the war as the most potent technical force in the metal industry.

And as we look to the future, in the service of more than 20,000 members, your Society has a greater opportunity than ever before and a greater challenge. And I believe it is ready for the task ahead.

The American Society for Metals has always been an educational organization. The articles of incorporation state that the principal purpose for which the Society was organized was for the exchange of information and the advancement of the knowledge of metals. While the object—education—has been predominant in all the activities of the Society, from that day to now, nevertheless the Board of Trustees during this past year has expressed a renewed interest in this important phase.

Their vision of possible increased educational activities to which the Society's tremendous resources might be directed consists in helping to plan metallurgical courses in the educational institutions of the country; the production of more educational and informational material for those now established in the metallurgical profession; and, in addition, an endeavor to provide greater knowledge on the subject of metals to those individuals who are contemplating engagement in, or now have some general interest or contact with, metal industry.

To assist in carrying out the first of these objectives the Board

appointed and has followed very closely the work of the A.S.M. advisory committee on metallurgical education under the chairmanship of Dr. Robert F. Mehl of Carnegie Institute of Technology. The agenda of this committee embraces a number of very important items and is evidence of the interest of A.S.M. in the training of future metallurgical engineers. It seems logical, also, that the A.S.M. should have this interest because within its membership is the largest group of metallurgists in the world.

In order to accomplish the second objective, a very active endeavor is being made to assemble and distribute a greater fund of metallurgical information than has ever previously been sent to the members. This includes the new and improved Review of Current Metallurgical Literature, a bibliographic service now appearing in the monthly METALS REVIEW, as well as immediate reporting of new materials and processes developed during the war and being made available now for release to civilian use. An enlarged and completely revised edition of the METALS HANDBOOK now in preparation should be a decided advantage for the members. The distribution is planned for this autumn. More than 35,000 copies of the present edition have been issued and are in use all over the world. Now, once again, some 2500 metal men are at work on additions and revisions that will make this 1800-page book the outstanding reference for information on metals and metal industry operations. And, in the same manner, your other regular publications are being expanded to handle all metallurgical engineering developments. Extensive plans are definitely in progress to revise the information in all the books published by the American Society for Metals and many additional books are contemplated to provide increased coverage and service to education.

The Society has pioneered in attempts to utilize visual methods of disseminating metallurgical information. The Educational Committee and Ohio State University in 1940 completed "Metal Crystals", a movie on the structure and growth of crystals and the construction of phase diagrams. It is hoped that future efforts will make it possible to actually visualize such difficult metallurgical phenomena as the mechanics of heat treatment, the hardening of steel, recrystallization and grain growth, etc. This assignment will tax the ingenuity of the combined metallurgical, educational and photographic fraternities.

Finally, in order to picture the metal industry to that large number of individuals embraced in that group from the senior in high

school to the noncollege graduate in the metal industry, there is planned for publication a number of Techbooks. These Techbooks are to be written clearly, interestingly and easily intelligible. It will be the first time that a series of practical Techbooks has been available for the practical man who wishes to know more about the methods or the materials with which he is or hopes soon to be working. These books, available at a very low price, will be helpful in creating an interest in the reader's work as well as a pride in his accomplishments. A number of unusual and new editorial publishing features will assist in the presentation of the text material in an easily understandable and self-educational style.

There has always been in the national headquarters a small library on the subject of metallurgy. However, it is now being expanded. Increased space is available in the property recently purchased by the Society, which is adjacent to the headquarters. The aim of the library expansion at the present time is to build up an active working library, specializing in assembling books and publications of recent date. It is not possible to have the library at maximum capacity and efficiency in a short period of time, but it is the ultimate expectation of the Trustees that there will be in time a very fine A.S.M. metallurgical library which will in many surprising ways be of assistance to the members wherever they may reside.

In short, the resources of your Society are at peak strength and it is ready to assume any services that will promote the arts and sciences connected with the production, fabrication, processing and use of metals.

During the past year, the Secretary and I visited 38 Chapters of the Society. We also had the pleasure of meeting with most of their Executive Committees. The mutual exchange of information and experiences was of benefit to all. The Chapters are indeed in a strong and alert condition—financially and spiritually. The Executive Committee members and Officers are unselfishly devoting countless hours to our primary objective—increased and improved service to members. I know that their efforts and plans will, as usual, receive the wholehearted support and co-operation of each member of the Society.

It has been a high honor to me to serve you during the past year as President and it is a most satisfying feeling to know that you are in such splendid hands under the leadership of the new president, Dr. Charles H. Herty, Jr., and the best technical society Secretary-Manager in America, W. H. "Bill" Eisenman.

TREASURER'S REPORT

HARRY D. MCKINNEY, *Treasurer*

The fiscal year which ended August 31, 1945, brought to conclusion another very successful year for the Society, and it affords your Treasurer a great deal of satisfaction to submit this report.

The cash balance on August 31, 1945, was $92,917.62 and includes no advance show receipts as has usually been the case in prior years.

Security holdings of the Society increased the past year from $622,911.43 to $848,534.11 or 36.2 per cent. Surplus funds permitted the purchase of additional securities which materially strengthened the security account.

Total assets increased from $1,002,394.65 to $1,155,724.43 or 15.3 per cent, but again it should be noted that no advance show receipts are included in the current figure. The true measure of the year's success is to be found in the increase of earned surplus from $742,644.82 to $974,289.51 or 30.8 per cent.

Accounts Receivable increased in total from $20,890.92 to $41,389.14, due largely to the increased volume of advertising in METAL PROGRESS. An average collection period of thirty (30) days is maintained and the receivable accounts are in a very healthy condition.

General Inventory has increased over $11,000.00 due to a reprinting of the METALS HANDBOOK. METAL PROGRESS, TRANSACTIONS, and METALS REVIEW inventories consist of stocks of paper, cover stock and envelopes used from month to month.

Office furniture and fixtures has increased over $5,000.00 due to the purchase of furnishings for the office building adjacent to the headquarters building.

Improvements made during the year to this newly acquired headquarters addition cost $9,600.00. Depreciation of approximately $3,000.00 was written off on buildings in accordance with the authorized amortization program.

The Finance Committee, together with the Trustees and other officers of the Society, are in constant touch with the A.S.M. fiscal agents, The Cleveland Trust Company, and a close contact with the investment market is maintained. Frequent conferences are held to consider investment problems.

Additional purchases of securities during the year have added $217,000.00 to government holdings and $23,000.00 to common

stocks, while disposals effected during the year have reduced bond holdings $19,000.00.

A study of our securities portfolio shows that approximately 62 per cent on the basis of cost is made up of government, utility, railroad, and industrial bonds, and the balance of approximately 38 per cent is made up of common and preferred stocks. Details of these investments will be shown in the published annual report.

The investment portfolio now shows:

Real Estate (Headquarters)	$ 71,960.00		7.9
U. S. Government Bonds	460,560.00		50.2
Canadian Government Bonds	15,640.00	1.7	
Canadian Utility Bonds	10,250.00	1.1	2.8
Public Utility Bonds	5,250.00	0.6	
Railroad Bonds	18,630.00	2.0	
Industrial & Miscellaneous Bonds	23,620.00	2.6	5.2
Common Stocks	280,460.00	30.6	
Preferred Stocks	30,310.00	3.3	33.9
	$916,680.00		100

As of August 31, 1945, the total cost of securities (excluding Headquarters) was $844,716.01 and the market value was approximately $917,188.00.

The net quick assets at August 31, 1945, amounted to approximately $958,700.00 or 1.6 times the total operating cost of the Society and the Cleveland Exposition during the year then ended.

The statement of income and expense for the past year is as follows in total:

Income	$863,594.10
Expense	631,949.41
Excess of Income over Expense	$231,644.69

The net cost to the Society for the publication of TRANSACTIONS was approximately $22,000.00.

The gross income from METALS REVIEW increased approximately $15,000.00. This gain was largely expended in increased production and promotion costs. The net cost to the Society for this publication was $14,700.00.

Miscellaneous items such as reprints, books purchased for resale, pins and buttons, automobile emblems, etc., balance out on approximately equal cost and income.

The sale of books published by the Society amounted to $38,940.20, and this represents a very nice activity for the Society and a real service to the members.

CONDENSED AUDITED BALANCE SHEET

As of August 31, 1945

ASSETS	
Cash	$ 92,917.62
Securities	848,534.11
Accounts Receivable	41,389.14
Inventories	51,090.38
Other Assets	12,767.59
Real Estate	71,961.01
Office Furniture, Fixtures and Equipment	17,151.64
Deferred Charges	19,912.94
	$1,155,724.43
LIABILITIES, RESERVES, AND SURPLUS	
Accounts Payable	$ 36,906.12
Reserves	135,000.00
Deferred Income	9,528.80
Surplus	974,289.51
	$1,155,724.43

INCOME AND EXPENSE STATEMENT

Year Ended August 31, 1945

INCOME			
METAL PROGRESS		$408,382.28	
1944 Convention—Cleveland		189,087.95	
Memberships		129,087.13	
METALS REVIEW		23,317.05	
METALS HANDBOOK		16,049.68	
BUYERS' GUIDE		12,145.00	
TRANSACTIONS		10,897.64	$788,966.73
Books published			38,940.20
Interest and dividends earned			24,501.31
General reprints and index			4,928.23
Discount earned			4,414.65
Sundry sales, etc.			1,698.27
Books purchased for resale			144.71
TOTAL INCOME			$863,594.10
EXPENSES			
METAL PROGRESS	$268,405.84		
1944 Convention—Cleveland	96,497.64		
General expenses	72,766.24		
METALS REVIEW	38,076.51		
TRANSACTIONS	33,013.78		
Books published	26,769.31		
Secretary's office	18,472.97		
Membership	17,816.65		
Accounting Department	12,182.50		
BUYERS' GUIDE	11,537.69		
Headquarters	9,166.44		
METALS HANDBOOK	6,354.47		
Trustees	2,853.44		
President's office	1,949.59		
National committees	1,736.69	$617,599.76	
Loss on disposal of securities		5,513.39	
General reprints and index		5,208.58	
Medals and awards		870.99	
Library salaries, supplies, and books		844.84	
Miscellaneous merchandise, etc.		775.10	
Lectures		520.00	
Medal fund expense		366.75	
Research and educational contributions		250.00	
TOTAL EXPENSES			631,949.41
NET INCOME			$231,644.69

In our opinion, the accompanying balance sheet and related statements of income and expense and surplus present fairly the position of the American Society for Metals at August 31, 1945, and the results of its operations for the year, in conformity with generally accepted accounting principles applied on a basis consistent with that of the preceding year.

ERNST & ERNST
Certified Public Accountants

The total net sales of the METALS HANDBOOK for the year amounted to $16,049.68, which clearly indicates the continued popularity of this very valuable publication of the Society.

The membership again moved to a new high during the year, as is evidenced by the rising curve reflected in the following table showing gross income from this source during the past eight years.

YEAR	GROSS INCOME	RATIO (1938 equaling 100)
1938	$103,600.00	100.0
1939	108,000.00	104.2
1940	115,500.00	111.5
1941	131,500.00	126.9
1942	150,100.00	144.9
1943	179,500.00	173.3
1944	193,800.00	187.1
1945	216,500.00	209.0

The gross income from membership during the current year represented 25 per cent of the total income, or 34 per cent of the cost of operating the Society. Of the gross income from membership $88,350.00 was returned to the chapters.

At the end of the 1944-45 fiscal year, the chapters reported total assets, in addition to those of the national office, of approximately $162,982.57, which is an increase in net worth of $29,176.57 over the previous year. This reserve guarantees a continuance of service, activity, and progress of our various local chapters.

Your Treasurer gratefully acknowledges the assistance of the Finance Committee, the Board of Trustees, our President, Dr. Kent R. Van Horn, Secretary William H. Eisenman, and our Assistant Treasurer, Mr. A. A. Hess.

ANNUAL REPORT OF THE SECRETARY

WILLIAM H. EISENMAN, *Secretary*

The American Society for Metals on October 1, 1945, had a total membership of 19,629. Of this number 18,052 or 91.9 per cent were the member classification; 1367 or 6.9 per cent were sustaining members; while 184 or 0.9 per cent were juniors. There are 26 honorary and founder members.

On October 1, 1944, the Society had a total membership of 18,468. The membership this year showed a gain of 6.2 per cent.

TRANSACTIONS

Since the last annual meeting of the Society, Vols. 34 and 35 of the TRANSACTIONS were published and distributed to the membership on request in March 1945. These two volumes total 1272 pages and constitute 46 articles with their discussions.

Vols. 34 and 35 contain all of the papers presented at the 1944 Convention held in Cleveland with their discussions together with other interim papers received during the year. Vol. 34 also contained a report of the Convention. The president, secretary and treasurer's annual report for 1944 and other current items of record were also included in that volume.

METAL PROGRESS

Statistics for METAL PROGRESS during the fiscal year of 1945 show that maximum production was achieved in almost all respects. Total number of editorial pages (674) approached the all-time peak (699 in 1942). Total number of pages of revenue advertising (2045) exceeded the previous maximum set the year before by 5.5 per cent. Circulation continually rose with the mounting membership. Using as a yardstick the report of Audit Bureau of Circulations, net paid circulation as of June 1945 was 20,410 copies, as compared with 18,774 copies just 1 year previous.

The outstanding excellence of METAL PROGRESS's covers has warranted innumerable expressions of praise from A.S.M. members (as well as from publishers of other technical magazines). Lest we fall into the error of thinking that they overshadow the contents of the magazine, I select a few editorial achievements—without implying any invidious comparisons with others not included in this necessarily brief list.

Articles on war production appeared quite regularly, such as descriptions of manufacturing methods for steel and brass cartridge cases, and of seamless gun tubes, and a 32-page supplement containing the metallurgical study of enemy matériel prepared by the War Metallurgy Committee.

A complete report on jet propulsion and gas turbines—one of the first, if not the first, such summary of the engineering and metallurgy involved to be published in the United States.

Numerous articles on the properties and uses of the National Emergency steels, which were such an important factor in the conservation of scarce alloys during the war.

A series of articles on the fundamental principles and practical operation of controlled carburizing processes and bright annealing, by Floyd E. Harris and others.

An account of methods (by C. A. Liedholm and associates) whereby the start and progress of transformation in alloy steels may be determined during rapid quenching, and several diagrams showing the transformation of alloy steels on continuous cooling.

A series of articles by Given Brewer and others on the methods of stress analysis, illustrating generally how obscure metallurgical failures can be diagnosed by this new science.

A leading article on an important new alloy of aluminum known as R-301, by T. L. Fritzlen and associates.

Metals Review

The editorial policy for the Metals Review inaugurated in February 1944 has been continued during the present year with only minor changes in format and style. The publication continues to be divided into three sections: The first editorial section contains news of chapter meetings, national Society activities, personal notes about A.S.M. members and other news-worthy items of interest to readers. The second section is the A.S.M. Review of Current Metal Literature, an annotated survey of engineering, scientific and industrial journals and books, which serves to keep A.S.M. members abreast of current developments in the metal industry as described in the literature. Advertising is also included in this section. The third section is devoted to brief descriptions of new products and catalogs announced by manufacturers serving the metal industry.

The 12 issues published between October 1944 and September 1945 have all carried either 20 or 24 pages, with the available space allotted approximately as follows:

	Per Cent
Reports of Chapter and National A.S.M. Activities and other news items	25.4
A.S.M. Review of Current Metal Literature	43.3
New Products and Catalogs	15.2
Advertising	16.1

It will be noted that the bulk of the space has been allotted to the Metal Literature Review, a feature which has become established as an important and valued service to A.S.M. members. This has not been done, however, at the expense of the space allotted to chapter and national A.S.M. activities, as can be seen by a com-

parison with figures for previous years before the new departments were added to the publication. Reports for previous years indicated that an average of 85 per cent of an 8-page issue of THE REVIEW was allotted to A.S.M. activities and the remainder to advertising. The present figure of 25 per cent of a 20 or 24-page issue provides about the same amount of space as was used in 85 per cent of an 8-page issue, bearing in mind the fact that the present page size is considerably larger than formerly.

METALS HANDBOOK

The personnel of METALS HANDBOOK Committee during the past year was as follows:

G. V. Luerssen, Chairman
A. J. Herzig
J. B. Johnson
J. J. Kanter
A. L. Kaye
C. H. Lorig
J. T. MacKenzie
H. W. McQuaid
H. S. Rawdon
H. L. Maxwell
C. W. Obert
J. H. Romann
Lyall Zickrick
J. Edward Donnellan, Secretary

Mr. J. H. Romann was appointed to the committee as the official representative of the American Society of Mechanical Engineers, bringing to four the number of societies co-operating with the A.S.M. in this important work and having representation on the METALS HANDBOOK Committee. The other societies are the American Institute of Mining and Metallurgical Engineers, The American Welding Society and the International Acetylene Association.

The general policy of deferring specific plans for publication of the next edition until the end of the war was continued. The work of organizing subcommittees was carried forward and at the end of the year most of these were complete and ready to begin active work with the cessation of the war in August.

A third printing of the 1939 edition of the METALS HANDBOOK has been necessary. The third printing (December 1944) consisted of 15,000 copies, bringing to 50,000 the total number of copies of the current edition. These figures emphasize the importance of METALS HANDBOOK as a reference work in the metallurgical field.

Effective at the close of the year, Mr. J. Edward Donnellan resigned as Secretary of the METALS HANDBOOK Committee after more than 21 years of continuous service. Mr. Donnellan was the permanent Secretary of the Handbook Committee, and of its predecessor, the Recommended Practices Committee of the A.S.S.T., from the inception of this work until August of 1945. During this

time six editions of the METALS HANDBOOK were published. The 1939 edition stands as a monument to the faithful and outstanding service of J. Edward Donnellan.

Books

During the past fiscal year more than 12,000 books of all titles published by the A.S.M. were sold to members and others, including copies of the METALS HANDBOOK that were sold to nonmembers. During this period no new book titles were published by the Society. There have been many of our previous titles reprinted in order to take care of the demand for these technical books during the war period.

The "Engineering Alloys" book recently came off the press in its revised and enlarged edition. The number of alloys listed in this new edition exceeds 12,500.

Publications Committee

The Publications Committee for the year 1944-45 was made up of the following personnel: Clair Upthegrove, Chairman; R. M. Brick, R. E. Cramer, H. I. Dixon, A. E. Focke, W. R. Frazer, R. P. Koehring, R. W. Roush, A. O. Schaefer, G. A. Sellers, Gordon Sproule, R. L. Templin, T. S. Washburn, S. P. Watkins, B. B. Wescott, and Ray T. Bayless, Secretary.

During the year no formal meeting was held by that Committee. One had been planned for July 1945, but due to the limitations on conventions and travel, the meeting was canceled. However, at a later date, when it was learned that a convention could be held early in 1946, the Chairman, Professor Clair Upthegrove, met with members of the Society staff for the purpose of scheduling the papers that are being presented at this convention.

During the year the Committee reviewed 40 papers, approved 33 and declined 7. Thirty of the approved papers are being presented at this convention.

Preprints

The thirty papers being presented at this Convention were prepared in preprint form and distributed to those members of the Society who had requested them. The total number of pages for the 1945 preprints is 873. A total of 35,000 preprint copies were distributed free to the membership to date.

Educational Committee

The Educational Committee for the year 1945 was composed of the following personnel: C. W. Mason, Chairman; C. R. Austin, J. L. Bray, O. W. Ellis, J. O. Lord, W. D. McMillan, C. H. Shapiro, C. G. Stephens, and Ray T. Bayless, Secretary.

The Committee held one meeting on December 8, 1944, at National Headquarters. The main purpose of this meeting was the planning and selecting of subjects and authors for the four lecture series that are being presented at this Convention, on Thursday and Friday. The titles of the lectures are:

I—*Magnesium*—Chairman, Kent R. Van Horn
1. *Magnesium—Extractive Metallurgy*—By L. M. Pidgeon, University of Toronto
2. *Magnesium Structural Design*—By J. C. Mathes, The Dow Chemical Co.
3. *Magnesium Castings*—By N. E. Woldman, Bendix Aviation Corp.
4. *A Survey of Wrought Magnesium Alloy Fabrication*—By J. V. Winkler, The Dow Chemical Co.
5. *Corrosion and Protection of Magnesium*—By W. S. Loose, The Dow Chemical Co.

II—*Induction Heating*—Chairman, C. H. Shapiro
1. *Principles and Theory of High Frequency Heating*—By H. B. Osborn, Jr., Ohio Crankshaft Co.
2. *Induction Heating Circuits and Frequency Generation*—By P. H. Brace, Westinghouse Electric Corp.
3. *Practical Applications of the Motor Generator Type of Induction Heating (Frequencies up to 10,000 Cycles)*—By W. G. Johnson, Caterpillar Tractor Co.
4. *Practical Applications of High Frequency Induction Heating (100,000 Cycles and Up)*—By J. W. Cable, Induction Heating Corp.
5. *A Comparison of Induction Heating with Other Methods of Heat Treating*—By T. E. Eagan, Cooper-Bessemer Corp.

III—*Effect of Surface Stressing Metals on Endurance in Repeated Loadings*—Chairman, R. L. Templin
1. *The Problem Defined*—By H. F. Moore, University of Illinois.
2. *Measurement of Surface Stresses*—By W. M. Murray, Society for Experimental Stress Analysis.
3. *Methods of Applying and Tests Used Including Carburizing and Nitriding*—By J. O. Almen, General Motors Corp.
4. *Stressing Axles and Other Railroad Equipment by Cold Rolling*—By O. J. Horger, Timken Roller Bearing Co.
5. *Progressive Stress Damage*—By P. R. Kosting, Watertown Arsenal

IV—*Corrosion of Metals*—Chairman, J. L. Bray
1. *Basic Principles of Metallic Corrosion*—By C. W. Borgmann, University of Colorado
2. *Effect of Composition and Environment on Corrosion of Iron and Steel*—By C. P. Larrabee, Carnegie-Illinois Steel Corp.
3. *Corrosion Resistance of Stainless Steels and High Nickel Alloys*—By W. O. Binder, Union Carbide and Carbon Research Laboratories
4. *Corrosion of Light Metals (Aluminum and Magnesium)*— By E. H. Dix, Jr., Aluminum Company of America
5. *Copper and Copper Alloys in Corrosive Environments*—By H. L. Burghoff, Chase Brass & Copper Co.

Three of these lectures have been prepared in book form and will be available at the close of the lecture series.

The Committee considered several methods of visual education, among which was a demonstration of three-dimensional stereopticon projection as perfected by the Polaroid Corp. This method of visual education holds considerable promise and may be adopted in the preparation of lectures on the fundamentals of metallurgy.

Further consideration was given to expanding the scope of the "Metal Crystals" film which was prepared by this Committee 3 years ago, but war and post-war limitations has held this work back. So far this film has had 147 showings, among chapters, universities and industrial educational programs.

The Committee is keenly interested in the preparation of lectures on fundamentals of metallurgy by visual methods such as the sound slide projection and sound moving pictures.

New Chapters

From Sept. 1, 1944, to Sept. 1, 1945, charters were granted to: Cedar Rapids and Terre Haute

National Metal Congress and Exposition

Inasmuch as it has been customary to chronicle a statement relative to the National Metal Congress and Exposition which has usually been in progress at the time of the presentation of this annual report (October), we think it inadvisable to establish an exception to the rule. Consequently, we wish to report that the Twenty-seventh National Metal Congress and Exposition, which is now in progress, speaks valiantly for itself.

The technical program has not suffered in its postponement from the usual October time to the week of February 4. The educational lectures listed on pages 4 and 5 will be presented on Thursday and Friday, and have all the earmarks of quality and success.

The National Metal Exposition, with some 440 exhibitors, represents the largest ever held by the Society and at the same time establishes a national record as to the amount of space occupied by an industrial exposition.

This is the first exposition since 1941 which has not had some of its equipment and developments withheld because of secrecy. The timeliness of the Show is indicated by the tremendous number of exhibitors, by the outstanding attendance, and by the universal satisfaction expressed by all.

ANNUAL MEETING OF THE AMERICAN SOCIETY FOR METALS

Due to the necessity of postponing the Twenty-seventh annual convention to February 4 to 8, 1946, the 1945 annual meeting of the American Society for Metals was held in conjunction with the November 2, 1945, meeting of the Cleveland Chapter of the Society at the Cleveland Club. The entire membership of the Society was duly advised of this meeting by a suitable printed notice in the September 1945 issue of the METALS REVIEW.

The meeting was called to order by President Van Horn who stated that the first order of business was the President's, Treasurer's and Secretary's reports. In order to conserve time the reports were not read but were deferred for reading at the Twenty-seventh annual convention of the Society held in Cleveland, February 4 to 8, 1946. The reports appear in this volume of TRANSACTIONS.

ELECTION OF OFFICERS

PRESIDENT VAN HORN, *Presiding*: Complying with the constitution, I appointed on March 2, 1945, the following Nominating Committee, selected from the list of candidates suggested by eligible chapters prior to March 1, 1945:

Walter Crafts, Buffalo, Chairman

R. M. Brick, New Haven
C. E. Carlson, Los Angeles
P. C. Cunnick, Tri City
H. E. Handy, Boston
W. G. Hildorf, Canton-Massillon
H. K. Ihrig, Milwaukee
A. J. Smith, Cincinnati
L. C. Whitney, Pittsburgh

The committee met in Chicago on May 24, 1945, and made the following nominations for national officers:

FOR PRESIDENT

C. H. Herty, Jr., Bethlehem Steel Corporation, Bethlehem, Pa.—1 year.

FOR VICE-PRESIDENT

A. L. Boegehold, General Motors Corporation, Detroit, Michigan—1 year.

FOR TREASURER

H. K. Work, Jones & Laughlin Steel Company, Pittsburgh, Pa.—2 years.

FOR TRUSTEE

John Chipman, Massachusetts Institute of Technology, Boston, Massachusetts—2 years.

Walter Jominy, Dodge Chicago Plant, Chrysler Corporation, Chicago, Illinois—2 years.

A report of these nominations duly appeared in METALS REVIEW, June, 1945.

I have been informed by the Secretary that no additional nominations were received prior to July 15, 1945, for any of the vacancies occurring on the Board of Trustees. Consequently the nominations were closed. I now call upon the Secretary to carry out the provisions of the Constitution in respect to the election of officers.

W. H. EISENMAN, *Secretary:* Conforming with the provisions and requirements of the constitution of the American Society for Metals, I hereby cast the unanimous vote of the members for the election of the aforenamed candidates who were regularly nominated May 24, 1945.

President Van Horn then introduced the newly elected officers at which time President-Elect Herty said a few words of acceptance.

PRESIDENT VAN HORN: Has anyone present anything to bring before the meeting? If not, a motion to adjourn is in order.

SOCIETY ANNUAL DINNER

The annual dinner of the American Society for Metals was held on Thursday evening, February 7, 1946, in the Grand Ballroom of the Hotel Statler. More than 800 members and guests joined in this annual function of the Society.

Those persons seated at the speakers' table were: Harold Y. Hunsicker, Vice-Chairman, Cleveland Chapter, A.S.M., Aluminum Co. of America, Cleveland; Arthur E. Focke, Past-Trustee, A.S.M., Diamond Chain & Mfg. Co., Indianapolis; John Chipman, Trustee, A.S.M., Massachusetts Institute of Technology; Walter E. Jominy, Trustee, A.S.M., Staff Engineer, Metallurgical Dept., Chrysler Corp., Detroit; M. Gensamer, Campbell Lecturer, Pennsylvania State College; J. H. Van Deventer, President and Editorial Director, *The Iron Age;* Wendell F. Hess, President, American Welding Society, Professor of Metallurgical Engineering and Head of Welding Laboratory, Rensselaer Polytechnic Institute, Troy, N. Y.; H. C. Erskine, Manager, Cleveland Works, Aluminum Co. of America; M. A. Grossmann, Past-President, A.S.M., Director of Research, Carnegie-Illinois Steel Corp., Pittsburgh; Earle C. Smith, Recipient, The Gold Medal, Chief Metallurgist, Republic Steel Corp., Cleveland; Charles H. Herty, Jr., Incoming President, A.S.M., Assistant to Vice-President, Bethlehem Steel Co., Bethlehem, Pa.; Kent R. Van Horn, Out-

going President, A.S.M., Manager, Cleveland Research Division, Aluminum Co. of America; Gerald Wendt, Speaker, Editorial Director, *Science Illustrated*; Zay Jeffries, Past-President, and Honorary Member, A.S.M., Vice-President, General Electric Co. (received medal for Gerard Swope); Robert S. Archer, Past-President, A.S.M., and Recipient of Albert Sauveur Achievement Award, Climax Molybdenum Co., New York City; Alfred L. Boegehold, Vice-President, A.S.M., Head, Metallurgy Dept., Research Laboratories Div., General Motors Corp., Detroit; Louis Seltzer, Editor, *The Cleveland Press*; William P. Woodside, Past-President and Founder-Member, A.S.M., Chairman of the Board, Park Chemical Co., Detroit; Earl S. Shaner, Editor-in-Chief, *Steel,* Cleveland; Morris Cohen, Recipient, Howe Medal, Associate Professor of Physical Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.; Lewis S. Bergen, Trustee, A.S.M., President, Bergen Precision Castings, Inc., Pleasantville, N. Y.; Stewart G. Fletcher, Recipient, Howe Medal, Latrobe Electric Steel Co., Latrobe, Pa.; Stewart M. Grant, Secretary, Cleveland Chapter, A.S.M., W. S. Tyler Co., Cleveland; W. H. Eisenman, Secretary, A.S.M.

Presentation of President's Medal

The annual presentation of the President's Medal was made by Kent R. Van Horn to Marcus A. Grossmann, the twenty-fourth president, who served the Society so ably in 1944.

Presentation of Howe Medal

In honor of Dr. Henry Marion Howe, the distinguished scientist, often called the dean of American metallurgists, the Board of Trustees in 1922 established the first of its medals. The rules governing the award of this medal make the provision that it be awarded to the author or authors of the paper judged of highest merit, presented before the A.S.M. and published during any one year in the TRANSACTIONS of the Society.

The 1945 medal was awarded to the three authors of the paper entitled "Structural Changes During the Tempering of High Carbon Steel" which was published in Vol. 32 of TRANSACTIONS, 1944, page 290. The authors who were honored are Dara P. Antia, Stewart G. Fletcher and Morris Cohen. Each of these authors was presented with a certificate, a gold medal and a bronze replica.

Albert Sauveur Achievement Award

In 1934 the Board of Trustees established an award consisting of a metal plaque and certificate in honor of Dr. Albert Sauveur, distinguished metallurgist and for many years an honorary member of the A.S.M. The purpose of this award is to recognize a metallurgical achievement which has stood the test of time and stimulated others along similar lines to the extent that a marked basic advance has been made in the metal arts and sciences. The 1945 candidate was Robert S. Archer, metallurgical assistant to the vice-president, Climax Molybdenum Co., who was presented by Dr. Zay Jeffries following the reading of the citation of Mr. Archer's accomplishments. Dr. Kent R. Van Horn then made the award. The citation is as follows:

"Some of Robert S. Archer's greatest contributions to the metal industry are in the art of fabricating, compounding, and heat treating of aluminum alloys for which the 1945 Sauveur Achievement Award has been made. One method of seeking an evaluation of his achievements in the aluminum industry might consist in an examination of the patents which bear his name and a determination of the relationship of that patent literature to current products and processes. An imposing number of patents from 1923 to 1934 bear Robert Archer's name either as sole inventor or as co-patentee.

"Three alloys described in Archer patents may be said to be the base upon which present day aluminum forging industry has been built. These alloys are known in the industry as 25S, 51S and 14S. The earliest efforts to develop an aluminum forging industry were centered about the utilization of Duralumin, the only heat treatable wrought alloy then in existence. This complex composition contains copper, magnesium, and manganese as the principal alloying ingredients. In the early stages of the art of forging aluminum alloys, this composition was so difficult to handle that only small forgings could be made and these left much to be desired both in cost and quality. Archer recognized the need of aluminum-base alloys with better forging characteristics and, after extensive research, the alloys 25S and 51S resulted. The first of these, containing copper and manganese but no magnesium, is much more forgeable than Duralumin and with proper heat treatment develops comparable mechanical properties. 51S is an entirely new type of

heat treatable wrought aluminum alloy. It contains no copper, which is one of the important characteristics of the Duralumin type alloy, and the total amount of alloying elements is less than 2 per cent. This material is characterized by a forgeability approaching that of pure aluminum, but its mechanical properties can be so enhanced by heat treatment that it is in the "strong" alloy class. A number of magnesium-silicide alloys similar to 51S, developed subsequently, emphasize the superior combinations of workability, strength, and resistance to corrosion available in this group of compositions. These two commercially forgeable alloys, 25S and 51S, served to open the entire field of aluminum forgings to commercial exploitation. Archer also contributed to the development of a modification of Duralumin of relatively low magnesium content (14S) which has definitely superior forging characteristics and which is capable of being produced with mechanical properties appreciably higher than those of Duralumin.

"The importance of these alloy developments to the aluminum forging industry, even at the present time, is illustrated by the fact that over 90 per cent of the aluminum airplane propellers being currently used by the United States forces are made of 25S alloy. At least one-half of all high output radial aircraft engine crankcases produced in recent years have been made from a 51S type of alloy. Millions of aircraft fittings have been made of 14S alloy and its superior hot workability is responsible for its recent expansion into the field of large extrusions for airframe structural members. The utilization of 14S alloy for this type of application made possible the meeting of aircraft schedules when extrusion capacities were critical. As a further contribution to the aluminum forging industry, Archer had an important part in the development of 32S alloy which is used for a large portion of the aircraft pistons.

"It is obvious from these illustrations that the alloys, in the development of which Archer participated, are the fundamental base of the aluminum forging business. Other of Archer's achievements in the aluminum forging art are not quite so tangible but are none the less real. For example, Archer's contributions to the development of ingot making procedures are noteworthy.

"When Archer first became interested in the aluminum industry, the field for the application of aluminum castings was

limited by the relatively low strength and ductility of the alloys then available. The novel principle of the heat treatment of aluminum castings and the alloy compositions adaptable thereto are described by Archer in the patent and technical literature. By means of these alloys and treatments, far superior combinations of strength and ductility can be obtained which made possible the opening of new and wider fields to the application of aluminum castings. Practically all the aluminum castings used in aircraft today are heat treated according to the procedures described in the aforementioned literature.

"Pistons for internal combustion engines were made of aluminum before Archer's connection with the industry. The alloy in general use at that time contained about 10 per cent copper. Archer participated in the development of a new type of piston alloy in which silicon was, to a large extent, substituted for the copper of the earlier alloy. This composition known in commerce as A132 also had the desirable property of a relatively low coefficient of thermal expansion. The previously described 32S is a forging alloy of the A132 type modified for better hot working characteristics. The availability of these alloys of relatively low thermal expansivity served greatly to increase the application of aluminum pistons.

"As a result of his work with silicon as an alloying ingredient in aluminum, Archer early came to the conclusion that the excellent casting characteristics of the aluminum-silicon alloys should be of use in the die casting industry. His efforts, following this line of thought, were to a great extent responsible for the application of the silicon type alloys in die castings. It can be stated that the present aluminum die casting industry is based to no small degree upon the silicon type of alloy.

"In addition to Archer's achievements in the aluminum industry, his general interest in, and contributions to, the "Science of Metals" is evident in his co-authorship of the book by this title. His part in the development of a general hypothesis of precipitation hardening which served for years as an inspiration for much research work in this field of metal investigation is further evidence of his broad general interest in the metal industry.

"The Sauveur award is made for proven achievements in the metal industry. Archer's achievements are particularly evident at present in the greatly expanded aircraft industry. The

members of the armed forces operating the products of this industry, though generally unaware of the identity of the individuals contributing to the development of the materials of which they are composed, would undoubtedly agree that it was proper for us who are acquainted with such details fittingly to recognize the individuals concerned. Robert S. Archer is one of those individuals."

Conferring of the A.S.M. Medal for the Advancement of Research

The 1945 A.S.M. Medal for the Advancement of Research was awarded to Gerard Swope, retired president and director of the General Electric Co., in recognition of his consistent sponsorship, foresight, and influence in financing and prosecuting metallurgical research, which have helped substantially to advance the arts and sciences relating to metals. Due to conditions beyond his control, Mr. Swope, unable to be in attendance at this meeting, was represented by Dr. Zay Jeffries, who received the award for him.

In making the presentation, Dr. Kent R. Van Horn read the following citation that appears on the certificate of award:

"Gerard Swope of the General Electric Company has had a pre-eminent share in many of the important developments in metallurgy in recent years.

"Vested with authority as president of his company from 1922 to 1939 and as president during the war emergency from 1942 to 1944, he demonstrated his faith in the future by his unfailing support of a program of fundamental and applied research, in which metallurgy has played a predominant role.

"During his first long term of office, the appropriations for the Research Laboratory had no limitations other than its needs. Even in the depression years, the laboratory was maintained at a high level of efficiency. Throughout this period the scientists and engineers under his direction made important additions to the knowledge of metallurgy, and in this war emergency their contributions have been outstanding.

"Prominent among these developments are cemented carbides; carbon-free alloys comparable in hardness to hard steels; magnetic materials, soft and hard, metallic and nonmetallic; high temperature alloys; precision casting processes; hydrogen

brazing; resistance welding electrodes; special alloys for mercury turbines.

"His sustained interest in metallurgical research and his influence in providing financial support for its continuance have produced a profound effect on the metallurgical science of his time."

In receiving this award, Dr. Jeffries read Mr. Swope's prepared acceptance, which follows:

"On behalf of the many metallurgists in the organization of which I was president for about a score of years, I wish to express deep appreciation for this high honor. Of course I thought they were doing a good job but it is most gratifying to have this view so eloquently confirmed by the award of the Advancement of Research Medal by an impartial board of your great society.

"I am especially pleased that this is a medal for the support of research. We owe our high standard of living in the United States more to research and its application than to any other one factor. We here in the past forty or fifty years have found the magic formula for abundance. It has been gained the hard way with its path dotted with mistakes and tumbles along with the successes.

"The formula seems simple now. It is to perform and make use of research along a wide front to discover new truths leading to new and improved products, new and improved methods, better facilities for production and more and better jobs. These spell a high standard of living and better opportunities for individuals to enjoy life."

Conferring of Gold Medal of A.S.M.

The Gold Medal of the A.S.M., established in 1943, recognizes the recipient for outstanding metallurgical knowledge and great versatility in the application of science to the metal industry, as well as exceptional ability in the diagnosis and solution of diversified metallurgical problems. Dr. M. A. Grossmann presented Earle C. Smith, chief metallurgist, Republic Steel Corp., Cleveland, as the candidate for the 1945 award. In presenting him, Dr. Grossmann read the citation engrossed on the scroll which accompanies the medal. Dr. Kent

R. Van Horn then conferred the medal award upon Mr. Smith. The citation is as follows:

Citation for Mr. Earle Smith

"Earle Smith early recognized that a producer of quality steels should be able to deliver to each buyer a metal especially adapted to meet the requirements, not only of the part to be manufactured but also of the machining, fabricating and heat treating practices existing in that specific plant. To put this policy into effect he recruited, trained and directed a group of men known as contact metallurgists.

"In order to be able to produce steels of such diversified qualities he delved into the science of steel making—one notable study comprising the mineralogical constituents of basic open-hearth slags, and their influence on the steel bath as they progress from their immature, partly melted stages to the mature refining slags. He also applied the petrographic microscope to the identification of nonmetallic inclusions in steel.

"Because of his intimate contact with all phases of steel making he exercised a constructive and important influence on the development, interpretation and appreciation of the effects of grain size.

"His fundamental and extensive knowledge of the fabricator's needs and the steel maker's ability to meet them bore fruit during the stressful years immediately passed when his advice was sought on countless problems connected with ordnance steel. He was directly responsible for the successful rolling of ingots into blooms for shipment to distant gun forging plants.

"He is an illustrious example of the ability to discern the value of scientific findings and convert them to practical use to the advantage of steel producers and customers."

Address of the Evening

Dr. Gerald Wendt, editor of *Science Illustrated,* presented the main address of the evening entitled "The Atomic Age Opens".

STRENGTH AND DUCTILITY

By Maxwell Gensamer

IN this, the twentieth Campbell Memorial lecture, I propose to discuss that aspect of metallurgy which has come to be known as mechanical metallurgy. This kind of metallurgy comprises our efforts to understand the relationships among the measured mechanical properties of metals and their mechanical behavior in service, and to understand the way in which these mechanical properties and service characteristics are controlled by chemical composition and structure. This field was not one which received a large amount of consideration from Professor Campbell, but I am sure that if he were alive today it would command his attention. In his day, the factors that influence the *structure* of alloys engaged the talents of some of the best men, and Professor Campbell and his contemporaries nobly inaugurated the work which has given us such a complete understanding of how to control the structure of metals, that many of us now feel free to turn to other problems. Of course this does not mean that no problems in that field remain to be solved. Rather, progress there has been so great that it now seems profitable to turn to other fields. I have chosen to concentrate on the problem of trying to understand *quantitatively* how alloying elements in solid solution, and undissolved compounds distributed in these solid solutions, affect mechanical properties.

Broadly there are two aspects to what we mean by mechanical properties. These are succinctly called strength and ductility. By strength we mean the resistance of a substance to distortion or fracture, and by ductility we mean how much we may distort it before it fractures. There are many ways to measure these. I have chosen to concentrate on the simple tension test, for I believe it yields as much or more information than any other test. Indentation hardness, which has been widely used, yields much less information, and those who have used it *only* have missed much of interest. Even the tension test alone is inadequate to complete under-

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standing, but by and large other tests are useful to supplement what the tension test can tell us. This applies particularly to measurements of strength, for which the tension test is almost wholly adequate; but it does not apply so well to ductility, where the conditions of loading are much more important and much less well understood in their action.

There are two aspects to what we mean by strength. These

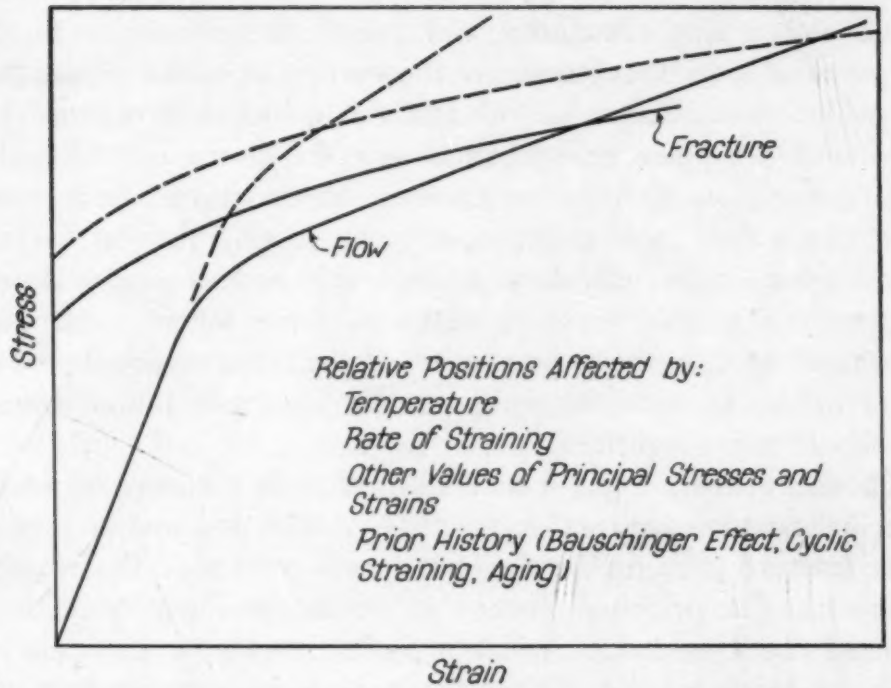

Fig. 1—Flow and Fracture Curves. Ductility at the point of fracture is determined by the intersection of the flow curve and the fracture curve. The course of both curves, and hence the local ductility, is affected by many variables including the variables listed.

are, first, resistance to flow, and second, resistance to fracture (1).[1] Both are functions of numerous variables, and can only be plotted on two-dimensional space by ignoring or holding constant all the variables but one. Both must be thought of as surfaces in multi-dimensional space. Among the variables which may influence both the flow and fracture stress, aside from composition and structure which are our principal concern today, we may list strain, time rate of straining, and temperature as the ones that most commonly are taken into consideration. In Fig. 1, the resistance to flow and the resistance to fracture are plotted against strain. I like to refer to these two curves simply as the flow curve and the fracture curve,

[1]The figures appearing in parentheses pertain to the references appended to this paper.

for the effect of deformation on strength is most familiar to us. Then the effect of the other variables is to shift these curves.

What is meant by stress and by strain in plots like this needs discussion. The state of stress at a point requires for its description, ignoring the orientation of the element of volume of the material under discussion, that we give the values of the three principal stresses. Only if we can find a function of the three principal stresses, by experiment invariable among the three principal stresses, can we plot a single quantity and call it the resistance to flow (or fracture) without reference to the particular values of the three principal stresses. Unless we can find this principal stress invariant, the ratio of the other two principal stresses to the one plotted as the resistance must be considered among the variables. And so with strain. As a first approximation, the von Mises criterion for flow (2), sometimes called the shear-strain-energy, comes pretty close to providing us with this stress invariant for flow; but it is not exact, depending at least on the average value of the three principal stresses. And as to the fracture strength, we still have not found even an approximate stress invariant.

There are many experimental difficulties in the way of finding a principal stress invariant for fracture, or for that matter measuring the fracture strength for any but special values of the variables affecting it. The principal stresses at fracture are usually unknown because of "necking-down", which precedes fracture in most circumstances for ductile metals. When the material is relatively brittle, so that fracture precedes necking-down, it is not too difficult to obtain a particular point on the fracture curve; it is that point where the fracture curve crosses the flow curve, which is when the metal breaks. Up to that time the metal is flowing because the resistance to fracture is less than the resistance to flow. And up until that time we cannot know where the fracture curve lies.

So, without changing any of the other variables, we can get only *one* point on the fracture strength curve. This so complicates the problem that no one as yet has found out how to determine the course of the fracture curve for the general case when both flow and fracture are affected by all the variables at our disposal. If we could find a variable that would affect the flow curve but *not* affect the fracture curve, then by changing *that* variable we could proceed along a series of flow curves to the breaking point on a *single* fracture curve, and so get enough points to plot it. It has been sug-

gested that temperature is such a variable, and that changing the temperature shifts the flow curve but not the fracture curve (3). I am afraid this is a vain hope, although it is true that the flow stress is affected more by changing the temperature than is the fracture stress.

Zener and Hollomon (4) have made a pretty good case for a single-valued functional relationship between speed of testing and temperature. If such an equivalence is borne out by more experience, then changing the speed of testing should be equivalent to changing the temperature, and tests carried out at different speeds should result in fracture at different limiting strains, that is, ductility should vary with the speed of testing, the way it does with temperature. Unfortunately, some of our experience with high speed testing in simple tension does not confirm this; high speed tensile tests on steel result in the same tensile reduction of area as slow speed tests.* Either speed and temperature are not equivalent, or what is more likely, resistance to flow and resistance to fracture are both affected, and in such a way as to yield the same deformation at fracture. But then, such high speed tests are open to questions of interpretation. Because of the limited speed of propagation of a plastic wave, it may be that the actual rate of deformation at the instant of fracture is independent of the speed with which the crossheads of the testing machine are moving apart, or at least that even very fast rates of testing fail to achieve a high rate of unit deformation at the point of fracture. This is one of the important and annoying complexities encountered in high speed testing.

We are, then, forced generally to ignore the fracture curve and to discuss the subject in terms of the flow curve and *either* the deformation or stress at fracture, which is unfortunate because it cannot give us the whole story. Under these circumstances it does not make any difference whether we talk ductility or cohesive strength, for these are the two co-ordinates of *a* point on the flow curve. But let us do the best we can, taking up first the flow curve, then ductility.

Our discussion of the effects of composition and structure on the flow strength of metals is to be based on the stress-strain curve obtained in tension. The very general applicability of tension curves to flow under other conditions of loading I shall refer to a little later. I think it advisable to discuss with you now the methods we

*Unpublished experiments by A. V. de Forest and Paul Shepler.

have used to analyze tension test data, methods which make the wide applicability of the tension test more apparent, and which have led to simplification of the relationships between composition or structure and the tensile properties. In Fig. 2 we have used data for a typical iron alloy. To the left the data are plotted in several ways,

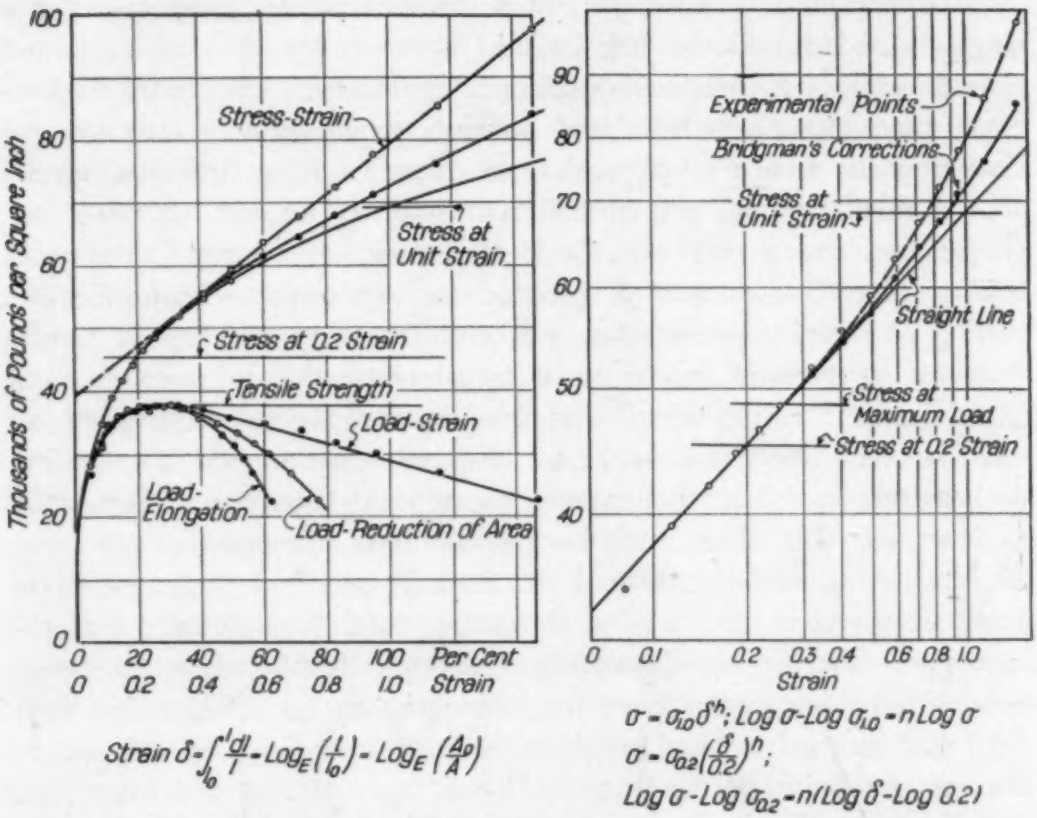

Fig. 2—Stress-Strain Curves Variously Plotted. To the left load-extension, load-reduction of area, load-true strain, average stress-average strain, corrected stress (Bridgman correction)-true strain, and extrapolated true stress-true strain curves are shown plotted directly. To the right the last three are shown plotted logarithmically. The slope of the log-log plot is the strain-hardening exponent; the stress for a strain of 0.2 has in this lecture been called the flow strength.

stress against strain or load against strain. To the right is a plot of the logarithm of the stress against the logarithm of the strain. It is our experience that, over the part of the stress-strain curve in which we may have confidence, the curve is always a straight line when plotted logarithmically. Beyond the point where necking-down begins, there are departures from the straight line plot for which we cannot yet make quantitative corrections; these departures result from the circumstance that we no longer have simple tension after necking-down begins, from the increase in speed of deformation when it is localized by this necking-down and possibly from some

reorientation effects. I think it only fair to assume that the curve would be straight to the point of fracture *if* we could avoid these disturbances.

Let us follow the plotting of a point on the final logarithmic curve, all the way from the original data to the final straight line plot. Necking-down begins at the maximum load (or maximum in load divided by original area, which is what is plotted here). If it were not for necking-down, it would be possible to go from a load or *nominal* stress versus extension curve, to a load versus reduction-of-area curve by calculation. Up to the time of necking-down, the one curve can be calculated from the other by using the fact that the volume or density of a metal is unchanged by plastic deformation, requiring constancy of the product of length by area. But after necking-down, this calculation is not possible, for the overall extension depends on the distribution of deformation.

To get a true stress-strain curve we must have the area, which must be measured after necking-down begins (5), (8). For each point then the diameter must be measured, at the point where it is least, at the center of the necked-down region. From the area we may calculate the strain, using always the logarithmic strain when the strains are large. The logarithmic strain takes into account the fact that the gage length is changing throughout the test, and it may be regarded as the true strain; it is simply the integral of dl/l, instead of $\Delta l/lo$, the conventional strain.

Dividing the load by the actual area gives the apparent true stress. This, however, is in need of correction because there are lateral tensile stresses in the necked-down region, which lead to higher average stress than would be the case if simple tension prevailed; and, unless the test is run very slowly, the rate of straining may be high. Bridgman (5) has attempted to calculate the correction required by the presence of lateral tensile stresses, but his correction is in our experience not quite enough; the corrected points still lie a little above the straight line in the log-log plot of stress versus true or logarithmic strain.

The fact that the log-log plot is a straight line means that all these tensile curves are general parabolas, fitting equations of the form $\sigma = \sigma_{1.0}\delta^n$, where σ is the true stress, $\sigma_{1.0}$ is the true stress at unit strain ($\delta = 1.0$), δ is the true strain (ln l/lo), and n, which we may call the strain-hardening exponent, is the slope of the log-log plot. It is interesting to note that the rate of strain hardening,

$d\sigma/d\delta$, obtained by differentiation of this equation, is equal to $\frac{\sigma}{\delta} \cdot n$, that is, the ratio of the stress to the strain multiplied by n; the greater n, the greater the rate of strain hardening. The constants which describe *completely* the whole course of the corrected stress-strain curve are then $\sigma_{1.0}$ and n.

We have chosen to discuss not $\sigma_{1.0}$, but the stress for a strain of 0.2, which we have called $\sigma_{0.2}$. The reason for this is that the $\sigma_{1.0}$ value is influenced very much by how the line is drawn through the experimental points, the only reliable ones of which center, for steel, around $\delta = 0.2$. We have used, generally, only points from $\delta = 0.1$ to $\delta = 0.3$, sometimes going a little lower than $\delta = 0.1$. Below $\delta = 0.1$, the points are quite reliable if the specimen is aligned very carefully before starting the test, and if a correction is made for the elastic strain. But above $\delta = 0.3$, necking introduces uncertainties. Extrapolation from $\delta = 0.3$ to $\delta = 1.0$ is too far under the circumstances. On the other hand, the stress at $\delta = 0.2$ is extremely reliable, and independent of the slope of the line drawn through the points. The equation for the curve can just as well be written

$$\log \sigma - \log \sigma_{0.2} = n \ (\log \delta - \log 0.2),$$

which is

$$\sigma = \sigma_{0.2} \ (\delta/0.2)^n.$$

Observe that the stress at the maximum load (which would be the tensile strength if the original area had been used instead of the actual area) is a point of no particular significance, for it does not come at a particular value of strain. The stress at $\delta = 0.2$ is the stress at the strain which is about average for the strain at maximum load, in iron about 20 per cent, but varying from 10 to 30 per cent. I shall call this stress at 0.2 strain the flow strength, in the same way that we have been calling the stress at a strain of 0.2 per cent or 0.002, the yield strength: 20 per cent strain, flow strength; 0.2 per cent strain, yield strength.

The general applicability of tensile data is now commonly recognized. The data are converted to the stress invariant by use of the von Mises function, and to the strain invariant required to satisfy the law of constancy of volume and the assumption of flow in the same way that a viscous fluid flows when acted upon by combined stresses (2). Neither of these methods of arriving at the stress

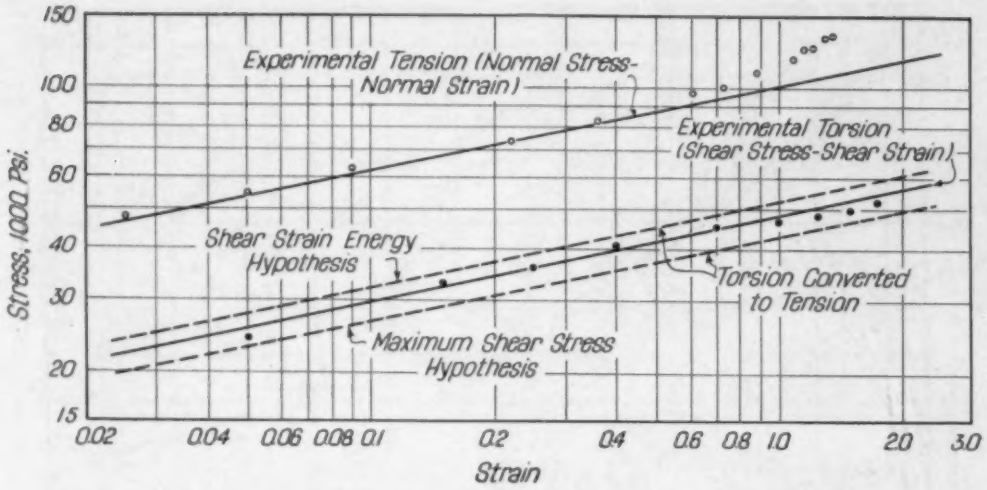

Fig. 3—Comparison of Tension and Torsion Data. The dashed lines are the tension data converted by calculation for comparison with the torsion data. Within narrow limits the agreement between data taken in tension and in torsion is quite good if tension data beyond the point of necking-down are neglected.

and strain invariants is exact, but they do yield values *almost* within the limits of experimental accuracy. Hence, for practical purposes, we may use the tension curve to calculate the curves for other loading conditions, where the *average* values of the three principal stresses are not too different. It would not do to calculate in this way a stress-strain curve for conditions including superimposed high hydrostatic pressure.

The tension curve may even be used to calculate the stress-strain curve in torsion, when the properly corrected curve is used. This is shown in Fig. 3, replotted from the work of Hollomon and Zener (4). The line through the tensile data, above, has been transferred by the von Mises or shear-strain-energy calculation, and again by the simple maximum-shear-stress hypothesis, to compare it with experimental data obtained in torsion, or twisting. Observe that, if we ignore the *uncorrected* points obtained after necking in tension, and assume that the data obtained in tension before serious necking apply to larger strains, there is good agreement between the tension data and the torsion data. It is the general experience, in all ranges of strain, that the torsion points fall between the values to be expected from the tension points by the application of the shear-strain-energy hypothesis and the maximum-shear-stress hypothesis. The spread between the two extremes is only 15 per cent.

Enough for the methods of analyzing the data. What have we learned about strength and composition, and strength and structure? In the first place, for all the steels we have tested, and we have

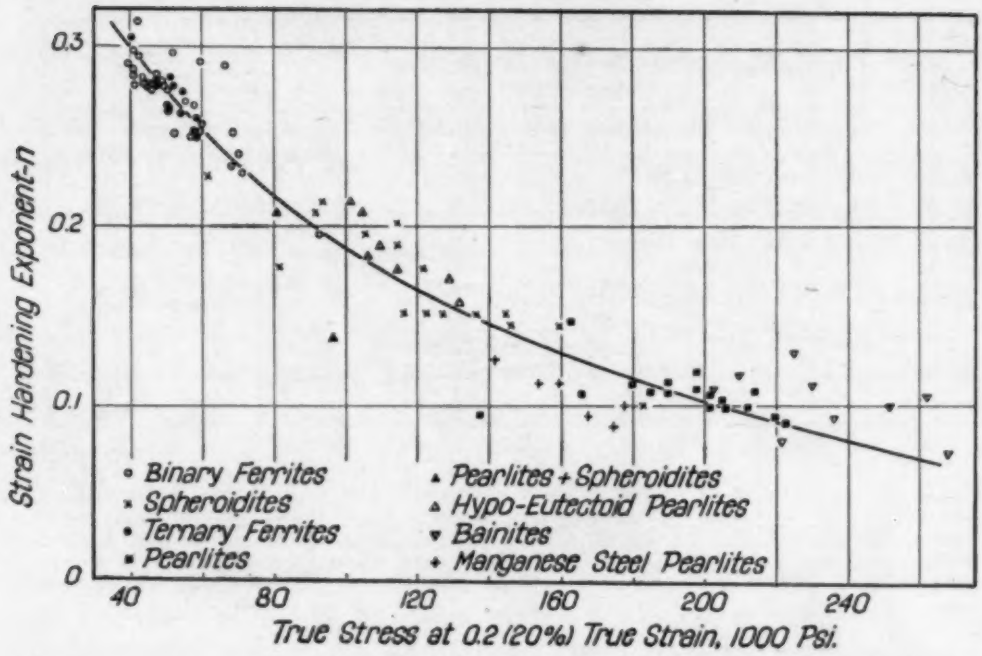

Fig. 4—Strain-Hardening Exponent and Flow Strength. This shows that in general the higher the strength of steels the lower the strain-hardening exponent, for a wide range of compositions and heat treated conditions.

worked only with steels, there is a fairly definite relationship between the flow strength and the strain-hardening exponent. This is shown in Fig. 4. The data are for decarburized alloys of iron containing from very little to almost the limit of solubility of many alloying elements, and for steels containing carbon up to the eutectoid, in many structural conditions including those typical of quenching and tempering as well as direct austenite decomposition. The general relationship is apparent. Whether the scatter is experimental or real and a result of variations in the alloys remains to be seen. It seems likely that we may expect variations of n at a given strength level. If n were really a single-valued function of the flow stress it would greatly simplify the task of discussing the effect of alloying elements and dispersoids on the flow strength of iron. We would need to report only the effects of those variables on the stress at a particular strain. The whole shape of the stress-strain curve would thereby be established, through this relationship between flow strength and the strain-hardening exponent. But I am afraid that this is true only as a first approximation, and that n must be determined.

When we first reported the data summarized here we used the usual true stress-strain curve, uncorrected beyond necking-down,

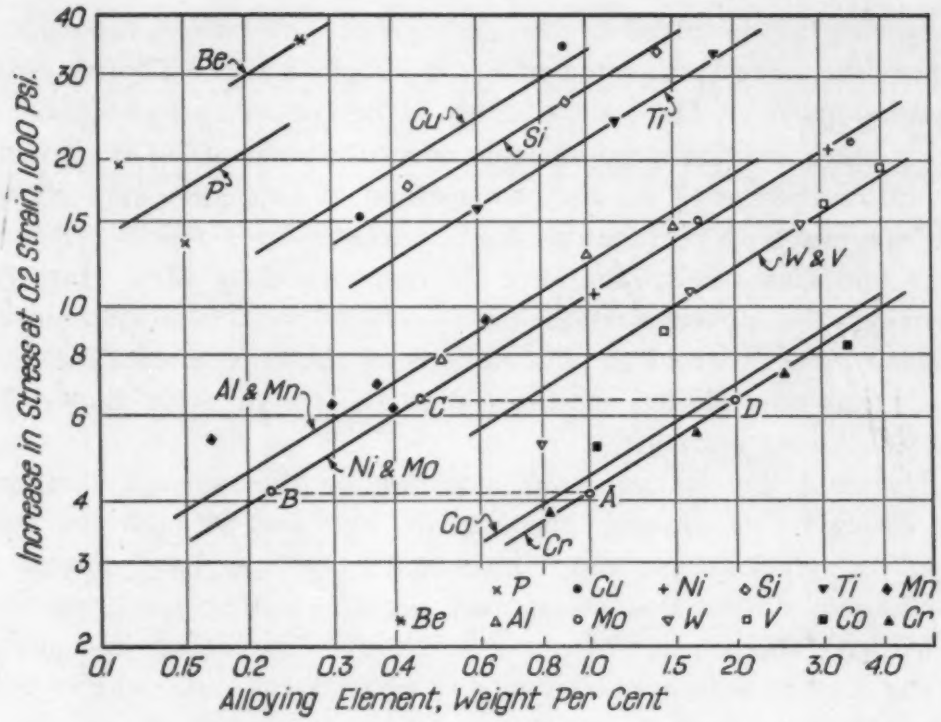

Fig. 5—Strengthening Effect of Alloying Elements in Iron. The nickel equivalent is the nickel content having the same effect on strength as 1 per cent of any particular alloying element, as at point A corresponding to point B for chromium above. The equivalent nickel concentration having a given amount of any alloying element is obtained by following across horizontally from the given concentration of alloying element to the nickel line; for 20 per cent chromium, corresponding to the point D above, the equivalent nickel concentration is given by the point C.

and *not* the log-log plot. Dr. Lacy and I (6) concluded that for alloyed ferrites, the slope of the usual curve increased with the intercept; whereas Mrs. Hartner (nee Pearsall) and Messrs. Smith, Low and Pellini and I (7) concluded for the heat treated carbon steels, especially the eutectoid, that the slope of the stress-strain curve was substantially constant, although not quite constant. These differences are now in part resolved. The variation is less at high strength levels, more at low. As a first approximation, all stress-strain curves for steel belong to the same family; as the strength level increases, the strain-hardening exponent and therefore the rate of strain hardening diminishes.

Now to tell you where we stand on what we started after. First, the effects of alloying elements dissolved in carbon-free ferrite. The recalculated data are compiled in Fig. 5. Some previously unreported results are here included with those of Dr. Lacy's. Frank Garofalo, working first at Carnegie Institute of Technology in the Metals Research Laboratory, and later at Pennsylvania State Col-

lege, with steels provided by the Carnegie-Illinois Steel Corporation, has provided some data on the effects of copper and phosphorus, notable omissions from Dr. Lacy's series. The copper and phosphorus-bearing steels contained appreciable concentrations of other alloying elements, necessitating the use of a method of successive approximations, starting with correlations based on Dr. Lacy's results. All of Lacy's and Garofalo's data were then reworked by Mrs. Hartner, improving the approximation until further corrections caused changes of the order of the fluctuations of the experimental results. The data as now plotted are thought to be free of error caused by residual alloying elements.

I should like to call your attention to the courage that we have displayed in drawing the line as we have through the two points for phosphorus; most investigators would have drawn a line of quite different slope, in fact negative rather than positive. But we have the courage of our convictions! We refuse to believe that the way to soften steel is to add phosphorus. A straight line of a given slope through a single point, as for beryllium, is also a good trick. And maybe all the copper in the higher copper alloy is not in solution, or the solution is supersaturated; if so, copper is not as effective as we think.

We have found that the effects of dissolved alloying elements are additive in a simple way. To calculate the strength to be expected in alloys containing two or more alloying elements, from data obtained with simple binary alloys, we have used the concept of equivalent-nickel concentration. Nickel is soluble over a wide range of concentrations and has a moderate effect on strength. To obtain the nickel equivalent of another element, simply observe from Fig. 5 what concentration of nickel will increase the strength the same amount as 1 per cent of the alloying element in question. For example, 1 per cent of chromium (point *A*) is equivalent to 0.22 per cent of nickel (point *B*). Then to express the concentration of chromium in an alloy as the equivalent nickel concentration, multiply the amount present by 0.22; for a 2 per cent chromium alloy this would be 0.44 per cent. The same result is obtained by reading across from the chromium line in Fig. 5 to the nickel line; 0.44 per cent nickel (point *C*) strengthens iron by the same amount as 2 per cent chromium (point *D*).

One can make a pretty good estimate of the strength of ternary or higher alloys by adding the equivalent-nickel concentrations of

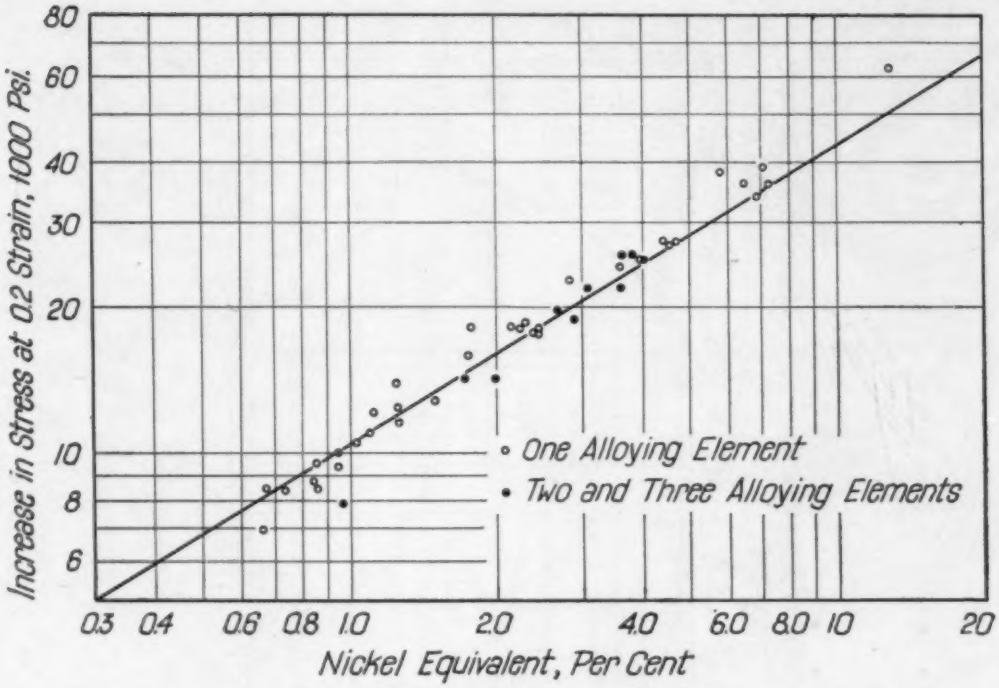

Fig. 6—Flow Strength and Equivalent-Nickel Concentration. Equivalent-nickel concentrations have been calculated by the method indicated in Fig. 5, adding together the equivalent-nickel concentrations of all the alloying elements present, including the residuals.

all the alloying elements present, and looking up the strength on the nickel line of Fig. 5. The filled-in circles of Fig. 6 fit the line about as well as do the points for the binary alloys. The line of Fig. 6 is the nickel line from the figure before this. In both figures the base strength for unalloyed ferrite has been taken as 30,000 psi.

The equivalent-nickel concentration is the quantity that we think most likely to be useful in efforts to understand why alloying elements have the effects they have. We have, of course, scratched our heads to find other properties of these elements that correlate well with their strengthening effect in iron. Solubility in iron correlates rather well (Fig. 7). By solubility we mean the extent of the primary α solid solution field in the binary phase diagram. Some of the data are not too certain. But in general, the relationship is good. There is a correlation with lattice distortion, observed by several investigators. Efforts to correlate equivalent-nickel concentration with ion or atom size do not give such a regular relationship, although there is a rough increase in strengthening effect with difference from iron in ion size.

Another kind of property, electrode potential, correlates pretty well, as is also shown in Fig. 7. Electrode and ionization poten-

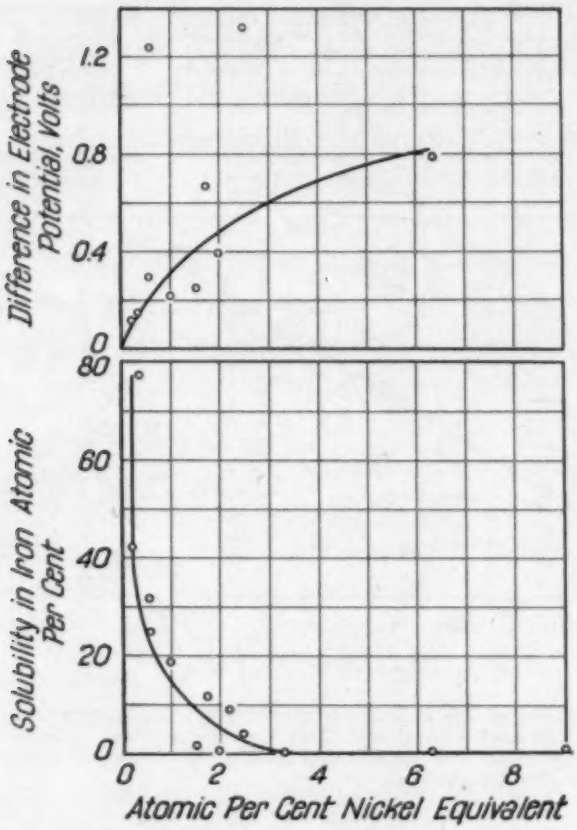

Fig. 7—Strengthening Effect, Solubility and Electrode Potential. Strengthening effect (nickel equivalent) vs. the solubility of the alloying element in iron, and the electrode potential of the pure alloying element in a cell reaction.

tials measure the work required to remove an electron. Somebody may see in this kind of thing some possibility of quantitative calculation. The points that lie farthest away from the curve that we have drawn are for aluminum and for titanium, both of which are trivalent whereas the other elements plotted here are divalent in the cell reactions used. Fig. 8 shows that for the elements near iron in atomic number, the strengthening effect varies fairly regularly with atomic number, excepting manganese.

If you should be more interested in tensile strength than in stress at 0.2 strain, Fig. 9 shows how regular is the relationship between them. For a complex alloy, the tensile strength may be calculated by calculating the total equivalent-nickel concentration, looking up the stress at 0.2 strain, and then getting the tensile strength from this plot.

Now for the strengthening effects of hard particles dispersed

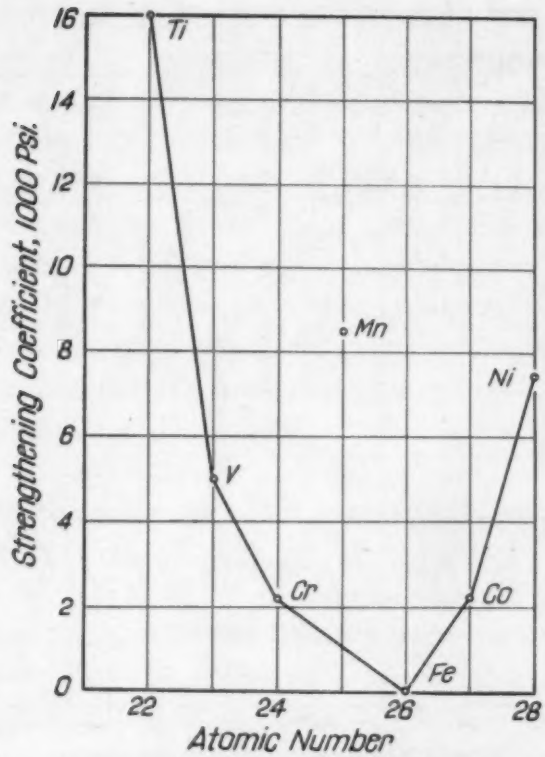

Fig. 8—Strengthening Coefficient and Atomic Number for Elements Near Iron in Atomic Number. The strengthening coefficient is the increase in stress at 0.2 strain caused by 1 atomic per cent of alloying element.

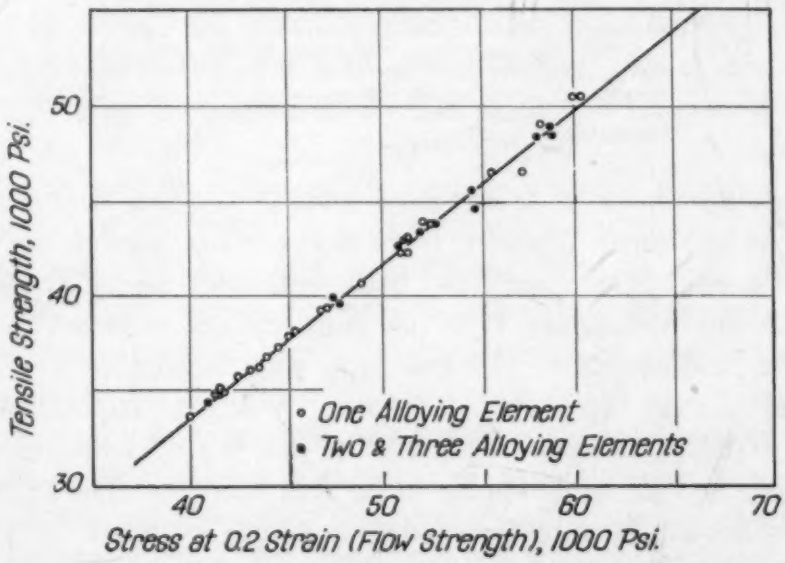

Fig. 9—Tensile Strength and Flow Strength. For steels there is a simple relationship between the flow strength and the tensile strength.

in iron. All the steels used in this study had almost the same alloy content, and from what we now know we would expect any effects caused by variations in alloy content to be quite negligible. Fig.

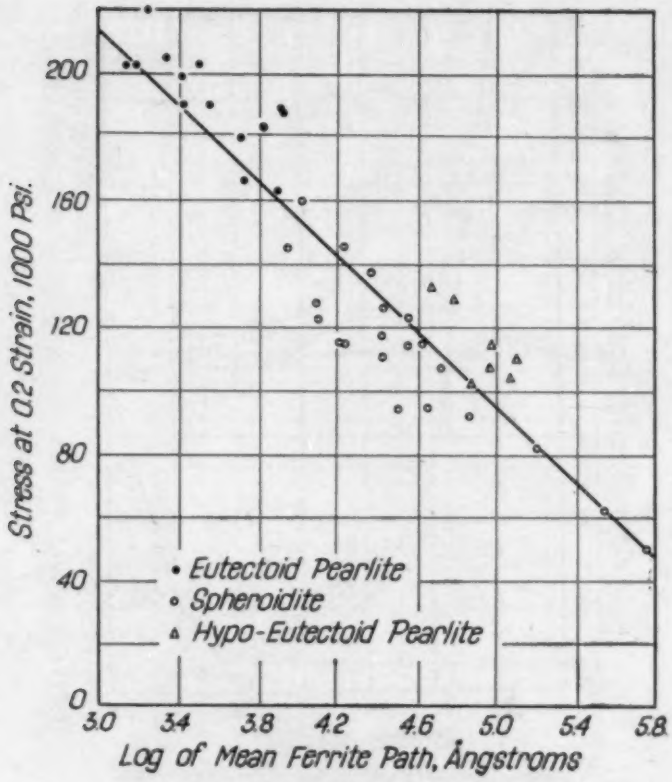

Fig. 10—Strength of Steels and Fineness of Microstructure. The flow strength is plotted against the logarithm of the average distance from one carbide particle to another through the ferrite. The hypoeutectoid steel contained about 0.40 per cent carbon; the eutectoid steels, approximately 0.80 per cent carbon. The hypoeutectoid steel in the pearlitic condition also contained ferrite, and the ferrite path for these specimens is the average for the entire specimen including the ferrite patches. The spheroidites were obtained by tempering following quenching.

10 is for the stress at 0.2 strain, and shows the dependence of strength on the mean distance from one carbide particle to another in these steels. The pearlites, both eutectoid and hypoeutectoid (0.78 per cent carbon and 0.38 per cent carbon, respectively), both seem to lie a little above the line; the spheroidites (quenched and tempered) on or below it. The line is drawn to minimize the squares of the deviations. The pearlite spacing measurements were quite tricky and are not wholly to be relied upon. The spheroidite measurements are straightforward (we simply counted the particles touching a number of lines on photomicrographs) but this method permits only the measurement of relatively coarse structures (7). For the hypoeutectoid steels, the ferrite path is the average for the whole alloy, across both ferrite and pearlite areas. *Strength* is not sensitive to *local* variations in structure; the alloy averages its strength over considerable volumes.

Observe that the amount of strengthening to be obtained in even coarse dispersions outweighs the strengthening effect of even large concentrations of alloying elements dissolved in ferrite. While it would be interesting to see how solution hardening and dispersion hardening add up when both effects are used in the same alloy, it is quite impractical to make the comparisons necessary because solution hardening is relatively so ineffective. The amount of strengthening to be expected by adding alloying elements to a carbon steel of fixed ferrite path would lie altogether within the scatter band of these experiments; going from a plain carbon steel to the highest alloyed of the S.A.E. steels might move the strength from the lower limit of the range to the upper limit. In such cases, statistical methods are necessary. In any event, we may safely conclude that alloying elements are useful to control reaction rates and carbide dispersion but have no value for strengtheners *as such.*

Perhaps the most important result to come from these studies is the observation that the shape of a carbide particle has no effect, provided the mean path from particle to particle through the ferrite be the same. Lamellar structures and spheroidal structures have the same strength for the same mean distance from one hard particle to another.

So much for strength; let us now discuss ductility. Fig. 11 attempts to show how ductility depends on gage length in 4 kinds of tests conducted on *sheets* of deep drawing grade steel. As in our discussion of strength, so for ductility it is necessary to define our terms. As said before, by ductility we mean the amount of deformation prior to fracture. If the deformation prior to fracture, all of it, were uniformly distributed in the machine part or test specimen, our problem would be much simpler. Even in the tension test, necking-down occurs because of nonuniform distribution of the deformation. Strain gradients exist. Because of these strain gradients, our measure of deformation depends on the gage length over which the strain is measured. *At* the neck, in a tension test, where the strain is greatest, the elongation or unit strain is *much* greater than at points remote from the center of the necked-down region. While the capacity of the material for deformation may be quite great, as measured by reduction of area at fracture or elongation *at* the point of fracture, the elongation over large gage lengths may be very little. Over somewhat shorter gage lengths it will be more, as the contribution of the elongation in the necked-down region is

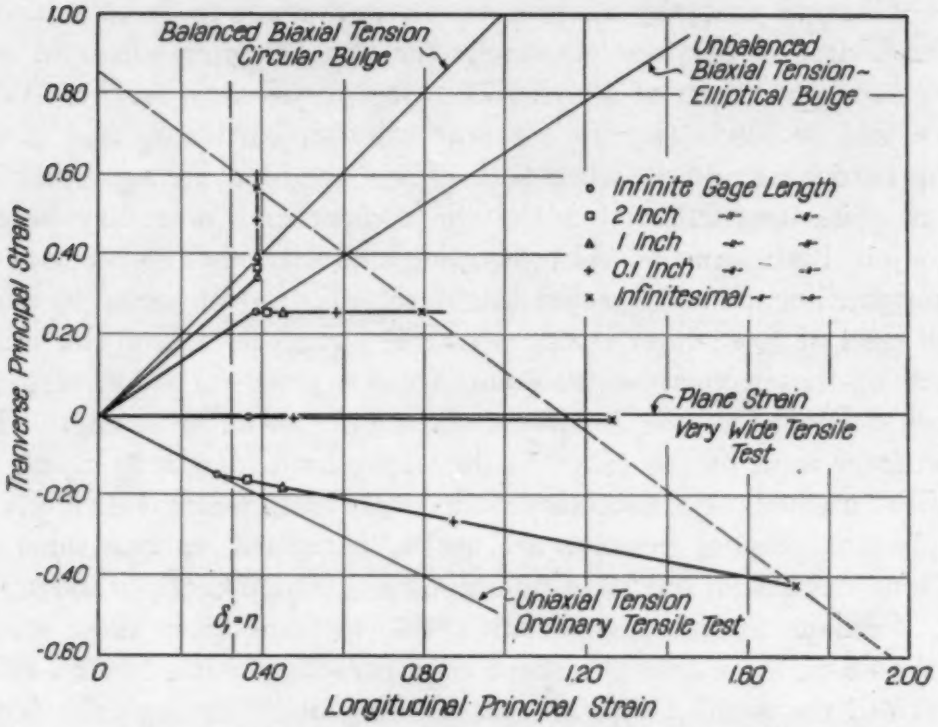

Fig. 11—Ductility in Biaxial Tensile Straining. Four kinds of tensile tests, ranging from the ordinary tension test for sheet metal to the balanced biaxial tension of the hydraulic bulge test. The material is deep drawing grade steel sheet. The deformation at the point of fracture varies with the ratio of the principal strains. The extension over a long gage length is practically independent of this ratio.

relatively a greater part of the whole when the gage length is short.

These are old and familiar facts, restated here to lead up to a circumstance that must be recognized. There are two aspects to the problem of describing ductility. First, we must know what capacity a metal has for deformation *locally;* what are the limiting strains, usually realized only in a very small volume adjacent to the point of fracture. Second, we must know how the deformation *distributes* itself, for this determines the total amount of deformation, which in turn determines the capacity of the machine part or specimen *as a whole* to absorb energy; and energy absorption is, in the last analysis, the important consideration — machines and structures are made of metals because they are capable of absorbing large amounts of energy without fracture.

These two aspects to ductility are quite separate and distinct things. One is determined by the cohesive strength, which is the force necessary to disrupt the structure and make a new pair of surfaces. The other has nothing to do with rupture, only with the distribution of deformation. This latter we may discuss in terms

of the strain gradient; if the strain gradient is large, the deformation will be local, and the energy absorption small; if the strain gradient is small, the deformation will be distributed, and the energy absorption large. Cohesive strength is probably a property quite distinct from flow strength and the factors that influence it; the strain gradient, however, is quite intimately, in fact entirely, determined by the plastic or viscous properties of the metal. Let us first take up the local capacity for deformation, as determined by cohesive strength, and then the distribution of deformation.

We may discuss the local capacity to deform in terms of *either* the true stress at fracture or the strain at fracture; given the flow curve, one determines the other and we may discuss the problem in terms of stress or in terms of strain, whichever is more convenient. Most of us are used to talking about ductility in terms of strain, and the local ductility in terms of the per cent reduction of area at fracture. In the simple tension test, as for a round bar, the per cent reduction of area is a complete statement of the limiting strain at fracture, for in the simple tension test, where the transverse strains in an isotropic metal are necessarily equal, and equal in sum to the longitudinal strain because of the requirement of constancy of volume, it is necessary to give only one of the principal strains. But under other conditions of flow, where the ratio of the principal strains throughout the test is other than $1 : \frac{1}{2} : \frac{1}{2}$ as in simple tension, it is necessary to specify the principal strain ratios as well as one of the principal strains. Further, the ductility depends on the values of all three principal stresses. It is now a commonplace that the ductility is greatly enhanced by high hydrostatic pressure superimposed on the deformation forces, and that metals are generally much more ductile in torsion than in tension.

To repeat, in Fig. 11 are plotted the values of two of the principal strains, measured over various gage lengths, for four different testing conditions. The material is deep drawing sheet steel, about 20 gage. At the bottom we have the ductility in the ordinary sheet metal tension test specimen, modified only in that the specimen is longer than usual to get away from the effects of the fillets, which are noticeable in the standard specimen. Now a sheet metal specimen, ½ inch wide by only 36 mils thick, does not continue to deform in simple tension once necking-down has begun. It contracts less in the width direction than in the thickness direction. Hence the lateral unit strain is less than ½ the longitudinal, the specimen

tending to remain at constant width after necking begins (but not achieving this, for the width does diminish somewhat). Hence fracture occurs, not for the ratio of the principal strains typical of pure tension, but at something between simple tension and plane strain.

The elongation over the infinite gage length is the strain at a point remote from the point of fracture, outside the necked-down region. If the gage length were infinite, so that the contribution from the larger elongation in the necked-down region were negligible, the average elongation over the whole infinite gage length would be the same as the elongation over a short gage length at this point remote from the necked-down region. The elongation over the infinitesimal gage length is calculated, by constancy of volume, from the reduction in width and thickness at the point of fracture; it is the same as the unit strain in the length direction measured over an exceeding short gage length.

Above the data for a sheet metal tension test in Fig. 11 are the data for the same sheet in a test in which change in the width direction is almost completely prevented by testing a specimen that is very wide (about 12 inches) and very short (only about 2 inches long). Such a specimen breaks without appreciable decrease in width. Hence the strain in the width direction is zero. Observe that the limiting strain in the length direction, which as in the tension test happened to be the rolling direction, is now considerably less than in simple tension (or as near to simple tension as we can come in a sheet test).

When we go to biaxial tensile straining, data for two varieties of which are shown along the other two lines in this figure, the maximum obtainable stretch is still less. The upper line, up to the point of necking-down, represents pretty close to balanced biaxial tension. It was obtained in the hydraulic bulge test, using a circular orifice. The sheet is clamped over a circular orifice and oil pumped into the chamber closed off by the sheet, blowing a bulge in the sheet. This bulge is substantially spherical; Fig. 12 is a photograph of such a bulge blown nearly to fracture. Over a fair sized area at the top of the bulge, the strain throughout the test remains uniform and the same in all directions in the sheet. This we call balanced biaxial tension. After necking-down begins, as in all such tests, balanced tension no longer prevails, for the necking-down region spreads as a trough, with all the stretch in one direction only.

Unbalanced biaxial tension may be achieved with control over the ratio of the principal strains, up to the point of necking-down, by blowing the bulge through an elliptical orifice. The stretch across the small dimension of the ellipse has to be more than across the long dimension. But after necking-down begins, again there is no more widening in the direction of the necked-down trough, and control of the ratio of the principal strains is lost. The same thing

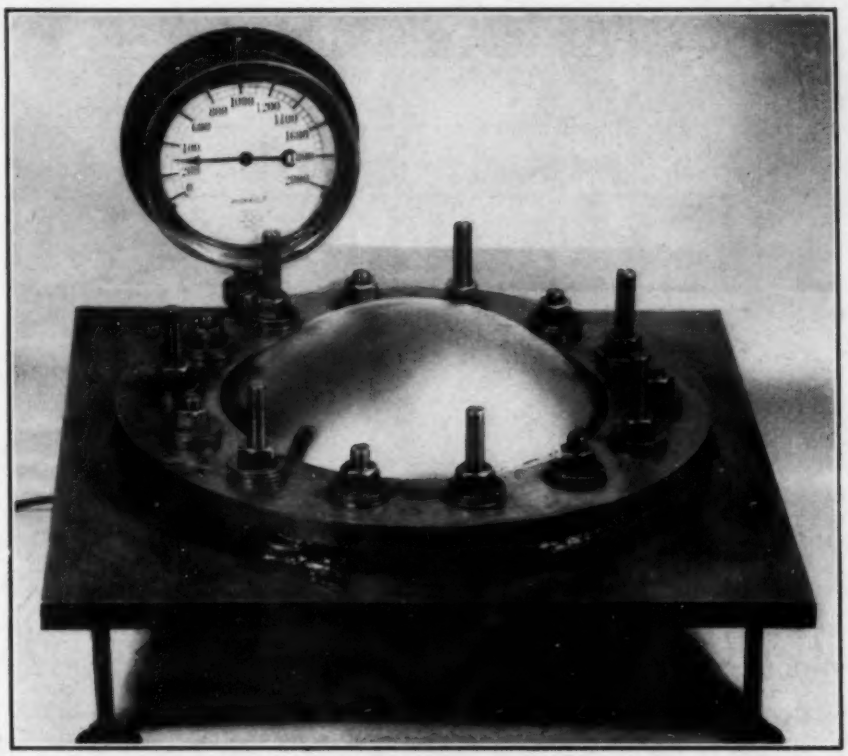

Fig. 12—Circular Bulge Test. A spherical bulge is blown through a circular orifice by hydraulic pressure. This produces balanced biaxial tension at the top of the bulge. For unbalanced biaxial tension, an elliptical orifice is used.

is true in tests of tubes under internal pressure, which have often been used; the advantages of these bulge tests over tube tests include the circumstance that necking-down occurs at larger values of strain than in tube tests. Of course, it is impossible to make satisfactory tubular specimens from sheet metal.

Observe that while the limiting deformation (the maximum local strain) varies with the kind of test, the uniform strain, or strain at points remote from the necking-down, depends very little on the test conditions. It is somewhat greater in the bulge tests than in simple tension. This fact may be more important than the deficiency of local ductility. The formability by stretching is fixed

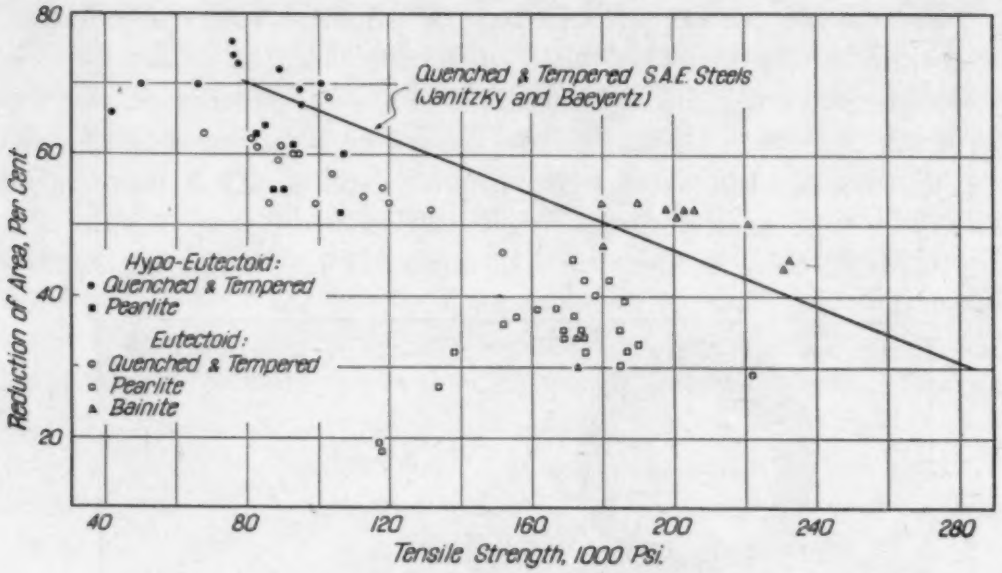

Fig. 13—Ductility and Strength. In general ductility diminishes quite regularly with increase in strength, but coarse pearlites are quite deficient in ductility. The extent to which temper brittleness may be responsible for this has not been established.

by the uniform or long-gage-length elongation; necking-down cannot be tolerated in a forming operation. Biaxial tension is detrimental to the local or fracture strain, but in a forming operation resembling this bulge test, more useful stretching can be obtained than in simple tension. It should be mentioned in passing that direct tension instead of bulging offers no similar advantage; the maximum uniform strain should be the same for all direct tensile stretching. The onset of necking-down is determined by geometrical considerations, which makes a bulge a more favorable shape than a flat sheet. Aside from this shape effect, necking should begin at the same value of the maximum tensile strain. Low and Hollomon have shown independently that this strain is numerically equal to the value of the strain-hardening exponent, n, discussed early in this lecture (8). Necking occurs when the rate of strain hardening drops to a critical value, and this occurs when the maximum principal strain equals n. Beyond this point, the sharpness of necking (the strain gradient) depends on other factors such as the rate of strain hardening and viscosity.

Before concluding this lecture with a more detailed discussion of the factors affecting the strain gradient, I wish I could discuss how the local or limiting deformation depends on composition and structure. But I am afraid that this is not possible. Cohesive strength is such a structure-sensitive property that very local condi-

tions obscure any larger effects. We all know how a very small amount of a brittle phase may utterly destroy the ductility of an otherwise very ductile metal. In steel, minute amounts of a precipitate may induce brittleness, as in strain aging and temper brittleness. For quenched and tempered steels free from temper brittleness there is a regular variation of ductility with strength (Fig. 13). While the ductility of quenched and tempered steels, as in Janitzky and Baeyertz's results (9), represented by the line in Fig. 13, varies regularly with strength, our work shows variation of ductility with type of structure. Whether the pearlites are really relatively brittle, as we might expect would be the effect of a brittle substance distributed in the form of plates continuous over fair sized areas, we cannot be sure; this may be a temper brittleness effect, for the production of such structures requires fairly long holding at temperatures at which it *might* be possible to develop temper brittleness. In any event, the bainites seem superior to the pearlites, and the two straddle the usual range for quenched and tempered steels. In the *particular* low alloy eutectoid steel used for this work, pearlite seems about as good as quenched and tempered structures, at high strengths, with bainite definitely better.

As to ductility in alloyed ferrites, we were not able to study ductility at all in the alloyed ferrite studies; variations in the degree of deoxidation were more important than the effects of the alloying elements. In the study of limiting, local deformations, then, very local conditions and minute amounts of embrittling phases are the important things. I think, though, that we have been paying a little too much attention to the limit of deformation at a point, and not enough to the distribution of deformation. I shall conclude by discussing the role of the strain gradient and the factors that influence its magnitude.

It is well known that when we change the temperature, or the speed, or the size and shape of the specimen, the amount of energy absorbed in a test to fracture changes greatly. Fig. 14, taken from Davidenkov (10), is typical. We could as well have plotted speed, and it might be that we could plot, could we choose the right parameters, the effect of size and shape.

Considering primarily the effect of temperature, we all know that the amount of energy absorbed varies from almost nothing at low temperatures to large values at high temperatures. The amount of energy absorbed is the sum of the energies absorbed in plastic

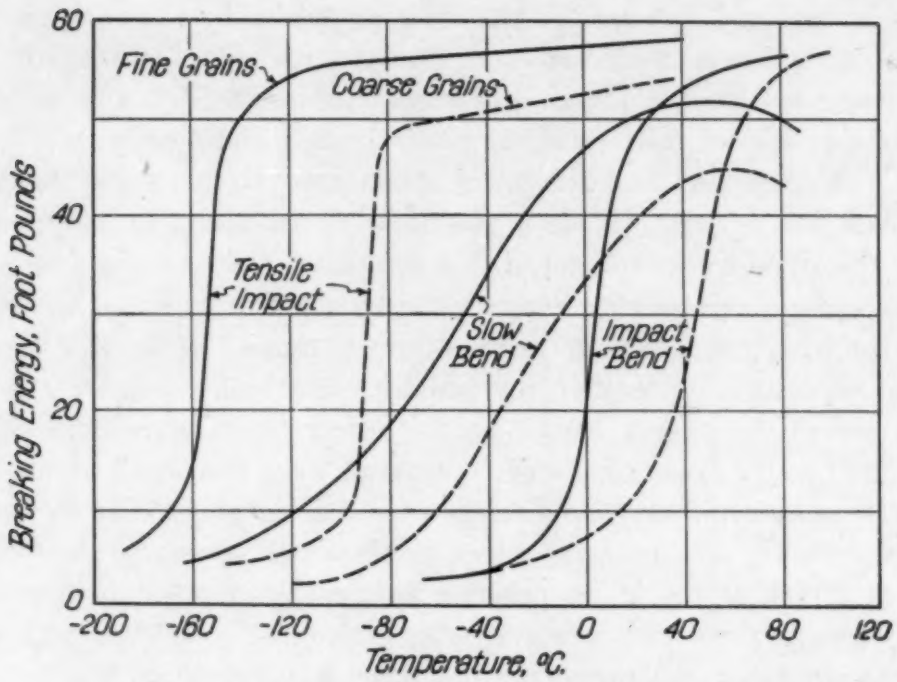

Fig. 14—Toughness and Temperature. Various factors influence the brittle temperature. Notches, coarse grains and high speed raise the brittle temperature. (After Dawidenkow and Wittman)

deformation by all the elements of volume in the test specimen. In each element, it is the area under the stress-strain curve up to the strain achieved by that element. Does the familiar result depicted in Fig. 14 mean that the limiting deformation or intrinsic ductility diminishes with temperature at the same rate as the energy absorption for the specimen as a whole? By no means. This great reduction in energy absorption results from extreme localization of the deformation at low temperatures. This localization of deformation becomes the controlling factor at a higher temperature in a fast test than in a slow one; and in a notch bend test than in a direct tension test.

Whether the local limiting deformation is great or little may be relatively unimportant; the value of the strain gradient may overshadow it in importance. To illustrate this point, consider this figure from MacGregor and Fisher's paper (11) on the impact and tension tests. Fig. 15 shows, as Moser had shown years ago but not as nicely as here shown by MacGregor and Fisher, that the volume of deformed metal in a V-notch Charpy impact specimen varies almost as much as the energy absorption. By deformed volume is meant the volume of metal which suffered a deformation in excess of 1 per cent. The energy absorption varies by a factor of over

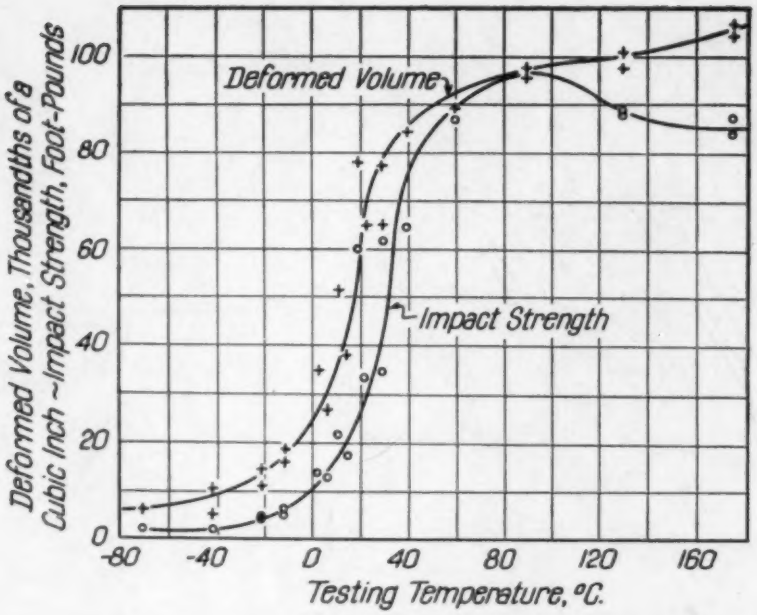

Fig. 15—Deformed Volume and Notched Toughness. The deformed volume varies almost as much as the notched toughness. (After MacGregor and Fisher)

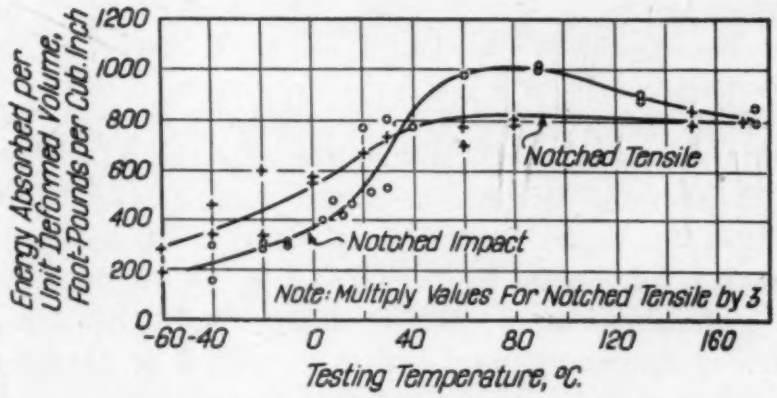

Fig. 16—Energy Absorption Per Unit of Deformed Volume. In notched-bar tests the energy absorbed per unit of deformed volume varies very much less than the total energy absorbed, because the deformed volume increases quite regularly with the total energy absorption. Low impact strengths are caused by concentration of the deformation into a small region. (After MacGregor and Fisher)

100 to 1, the deformed volume by a factor of 25 to 1. The energy absorbed per unit volume of deformed metal varies by a factor of only 4:1, as shown in Fig. 16. Locally, the material shows less energy absorption, but in nothing like the same degree as the specimen as a whole, because the deformation is localized. This is the principal effect of a notch and of changing speed and temperature. Fig. 17, obtained for me by John L. Fisher at Penn State College, shows this as hardness contours on broken Charpy bars of 3½

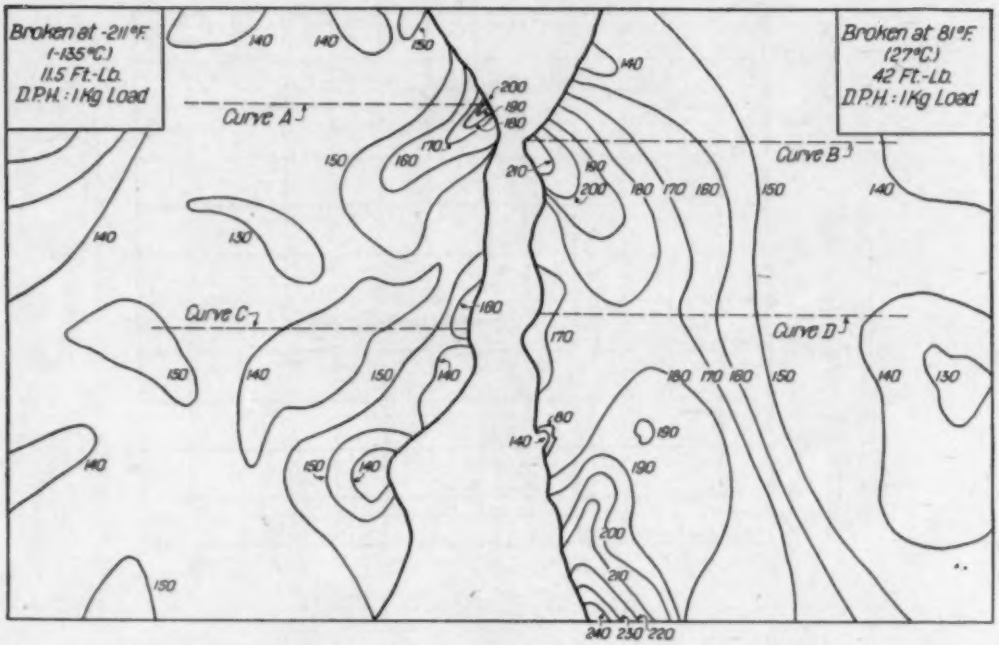

Fig. 17—Strain Distribution in Charpy Bars. Half of each specimen is shown. The bar to the left was broken at a low temperature and absorbed 11.5 foot-pounds. The bar to the right, of the same steel (low carbon, 3.5 per cent nickel), was broken at room temperature and absorbed 42 foot-pounds. The contours shown are actually indentation hardness contours. These can be correlated with the amount of strain. The variation of hardness along the dashed lines shown is plotted in Fig. 18. Data by John L. Fisher.

per cent nickel steel, broken at —211 degrees Fahr. and at room temperature. The lower temperature bar absorbed 11.5 foot-pounds against 42 foot-pounds for the higher temperature bar and yet shows almost the same maximum hardness (200 as against 210 Vickers with a 1-kilogram load). It can be shown that this means almost the same maximum deformation as the higher temperature bar. What is vastly different is the steeper strain gradient in the lower temperature specimen. Hardness traverses along the dashed lines of Fig. 17 are plotted in Fig. 18. The steeper strain gradient and the large but *localized* limiting deformation at the lower temperatures are apparent.

To understand how temperature, speed, size and shape affect ductility, we must study the factors that influence not only local deformation limits, but more important, the factors that affect this strain gradient. Consider a specimen in which there exists a stress gradient, as for the notched tensile bar depicted in Fig. 19. Here are shown the three principal stresses, longitudinal, tangential and radial, as well as the maximum shear stress which as a first approximation determines the rate of flow. This is for very deep hyperboloid notches, calculated by the method of Neuber by the staff of

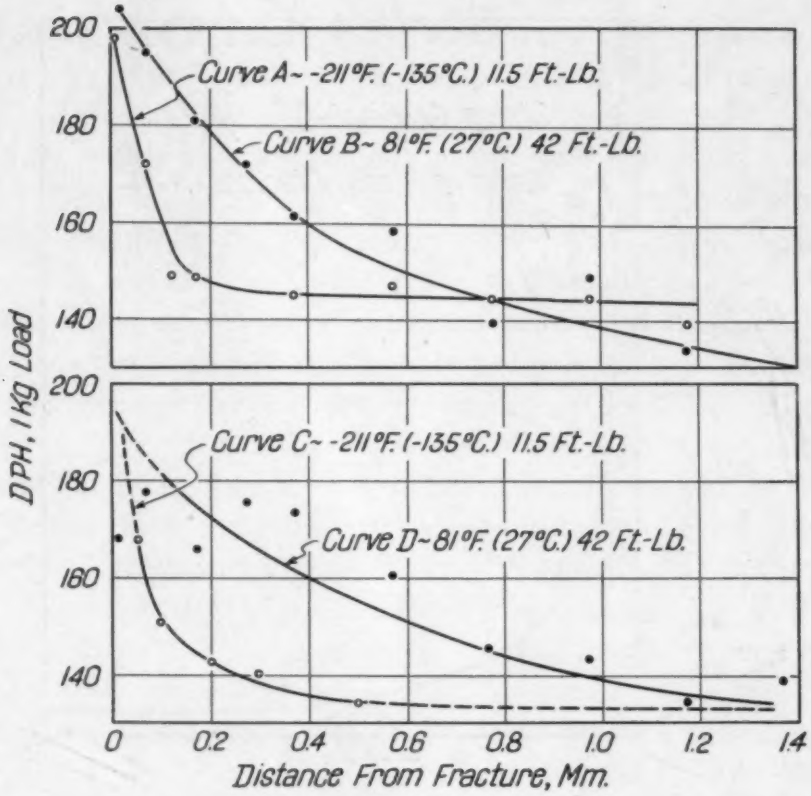

Fig. 18—Strain Gradient and Temperature. The deformation, indicated by hardness numbers, is spread to larger volumes when the energy absorption is high. For high energy absorption the strain gradient (slope of the deformation distribution curve) is low.

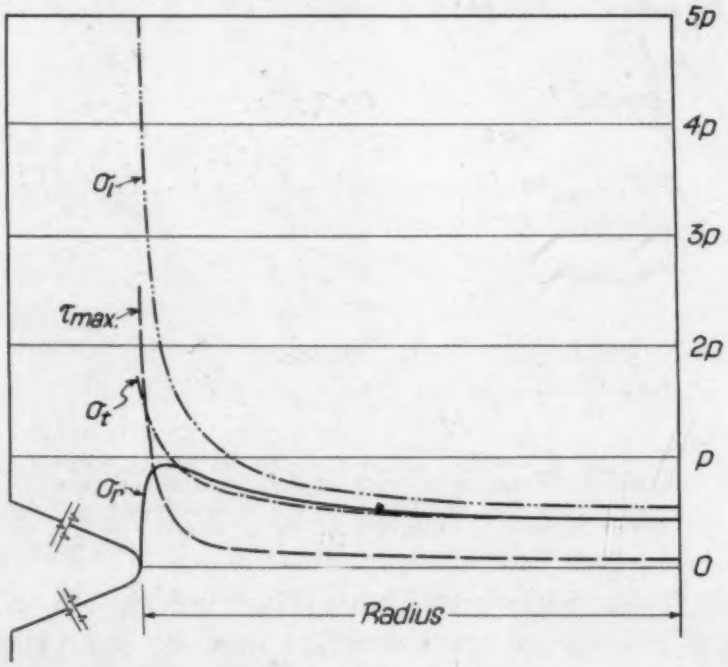

Fig. 19—Stress Distribution in Notched Tensile Bars. Variation, from bottom of notch to center line, of the three principal stresses and the maximum shearing stress. The average tensile stress is ρ.

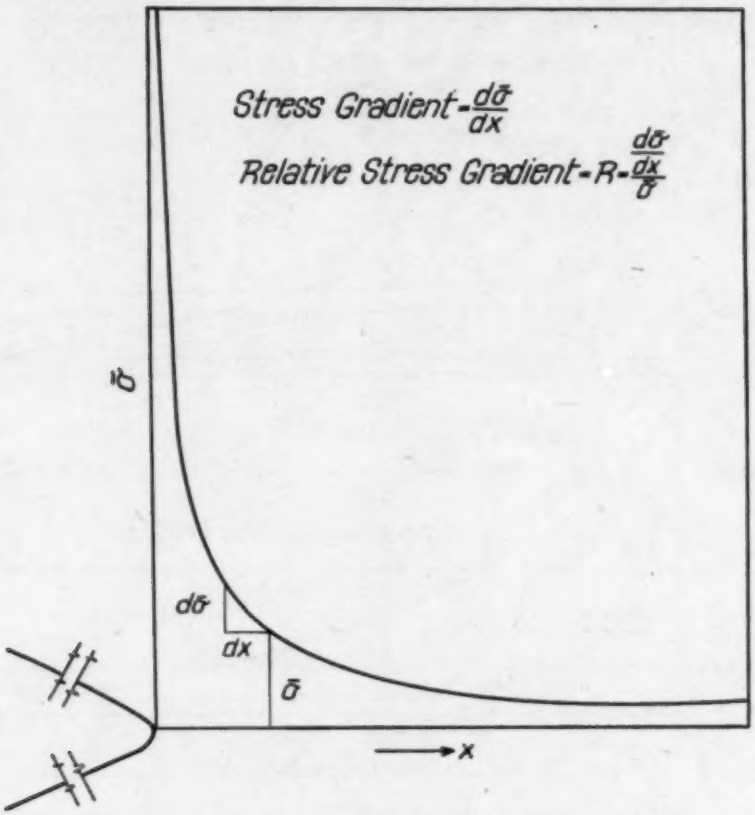

Fig. 20—Effective Stress Distribution in Notched Bars. The effective stress is proportional to the von Mises function of the principal stresses. The stress gradient at any point is the slope of the stress-distance curve at that point.

$$\sigma = \sigma_1 \delta^n \qquad \sigma = \eta \frac{\partial \delta}{\partial t}$$

$$\frac{d\sigma}{dx} = \sigma_1 n \delta^{n-1} \frac{d\delta}{dx} \qquad \frac{\partial \sigma}{\partial x} = \eta \frac{\partial}{\partial x}\left(\frac{\partial \delta}{\partial t}\right) = \eta \frac{\partial}{\partial t}\left(\frac{\partial \delta}{\partial x}\right)$$

$$\frac{d\sigma}{dx} = \sigma_1 \delta^n \frac{n}{\delta} \cdot \frac{d\delta}{dx} \qquad \frac{\partial \delta}{\partial x} = \frac{1}{\eta}\int_0^t \frac{\partial \sigma}{\partial x} \cdot dt + F(x)$$

$$\frac{d\delta}{dx} = \frac{\frac{d\sigma}{dx}}{\sigma} \cdot \frac{\delta}{n}$$

Fig. 21—Strain Gradient, Strain-Hardening Exponent, and Viscosity. Rate of strain hardening, and viscosity, should influence the strain distribution and hence the energy absorption in the structure.

the David Taylor Memorial Model Basin of the U. S. Navy at Carderock. The shear stress gradient may be quite high. Focus attention on this, or the von Mises flow stress function which is very similar as in Fig. 20.

We may define some of the variables that we shall discuss as follows: The stress gradient is the slope of the plot of stress against distance, $d\sigma/dx$. This divided by the absolute value of the stress gives the relative stress gradient, $d\sigma/dx/\sigma$. The same plot would do for the strain, δ, for which the strain gradient, in which we are interested, would be $d\delta/dx$. It can be shown that the *strain* gradient depends not on the capacity for deformation as measured by reduction of area or other principal strains at fracture, but on strain-hardening rates and viscosity. By viscosity we mean the coefficient relating the resistance to plastic deformation, to the speed of deformation. The following analysis may not be mathematically rigorous, but in essence it is sound and I hope convincing.

Consider first the effect of the strain-hardening exponent independently of the speed effect. Keep the speed of deformation constant. Consider only a particular point in the body, a point at the distance x from the surface, under stress σ, with a stress gradient $d\sigma/dx$. What should be the strain gradient? The steps in this argument are collected in Fig. 21. Let us assume the usual stress-strain curve, for which $\sigma = \sigma_{1.0}\delta^n$. If we differentiate this equation for the stress-strain curve with respect to x, we obtain, after some juggling, the result $\frac{d\delta}{dx} = \frac{\frac{d\sigma}{dx}}{\sigma} \cdot \frac{\delta}{n}$, which tells us that the strain gradient is greater for low n values and is less for high n values, and that it is proportional to the relative stress gradient. Of course, the stress gradient diminishes as the specimen deforms, so that the strain gradient may or may not diminish as deformation proceeds, depending on whether the increase in strain or the decrease in relative stress gradient proceeds at the faster rate; if the specimen did not break, the gradient would become zero, which means that eventually the strain would be uniformly distributed; of course this point is never reached. If this analysis be sound, what we want for widespread strain distribution and high energy absorption is a material tending to give a low strain gradient, and that means *a high strain-hardening coefficient.* The *plastic properties should determine energy absorption,* not only by determining the area under the stress-strain curve, but by controlling the distribution, or localization, of strain. This puts us back to something that can be studied as a function of composition and structure, with the factors that determine local ductility entering often in a minor way. Of course, a

drastic reduction of local ductility would be important, but where the ductility is fairly high, plastic properties may outweigh it in practical consideration.

Consider a viscous material. An ideally viscous material is one for which the resistance to flow is directly proportional to the speed of flowing. The best information we have indicates that steel at low temperatures (below the recrystallization temperature) is such a viscous material, if we discuss not the absolute stress, but the increase in stress above the static flow curve. But to keep the argument simple, let us talk about a simple viscous body, with no strain hardening. For such a body, the law of viscous flow states that $\sigma = \eta \frac{\partial \delta}{\partial t}$, where η is the coefficient of viscosity and $\partial\delta/\partial t$ is the speed of deformation. Differentiating this with respect to x, and manipulating, yields such an expression as this for strain gradient:

$$\frac{\partial \delta}{\partial x} = \frac{1}{\eta} \int_0^t \frac{\partial \sigma}{\partial x} \cdot dt + F(x)$$

A low strain gradient (and high energy absorption) requires a high viscosity. For steel, viscosity is much greater at high temperatures than at low. (Remember that viscosity as we have used it is the *slope* of the plot of stress against speed, and *not* the absolute stress.)

If the stress gradient is a function of the strain rate (and it can be argued that it must be for a viscous material) then the strain gradient must be a function of the speed of deformation and we would expect that increasing the testing speed would increase the stress gradient *and* the strain gradient, and thus would localize the deformation and reduce energy absorption. In this way, and possibly by affecting the strain-hardening exponent, speed effects are introduced.

Now I realize that the solution of these over-simplified differential equations, to calculate the distribution of strain and the energy absorption for a practical case would be impossibly difficult. But I think such considerations as these have value, for they point to the variables that must be measured if we are to explain and predict variations in notch sensitivity, size effects, temperature and speed sensitivity, etc. It is not now the practice to determine strain-hardening exponents, nor viscosity coefficients, as functions of speed and temperature. There are not enough data available to test the soundness of deductions such as these. But we can never hope for

complete control over the properties of metals in all applications until we succeed in reducing the qualities that must be controlled to a few simple, fundamental ones such as n and η. We know we are not *now* measuring the significant properties, for we cannot predict behavior in service to a satisfactory extent. Yet there are enough observable regularities to make us hope that such simplification may be possible.

I should like to summarize briefly what I have said, for the sake of emphasis.

1. Strength is not sensitive to local variations in composition and structure.

2. It depends in a very regular way on composition in solid solutions.

3. It depends in a very regular way on the mean spacing between particles in aggregate structures, and not at all on the shape of the particles except as this affects the mean spacing.

4. The effect of dissolved elements is insignificant in comparison to the strengthening effect of fine dispersions.

5. Ductility on the other hand is much affected by local conditions.

6. Local ductility, unless very low, may have little effect on energy absorption in a structure. The distribution of deformation may be of more consequence.

7. Efforts to improve energy absorption in structures by improving the local limiting deformation at the point of breaking are liable to be disappointing.

8. Improvement in deformation distribution by modification of the plastic properties of substances should be more profitable, when the intrinsic ductility is considerable. However, this should not be true at high strength levels, where intrinsic ductility is low enough to be a more important factor. It is still important for us to try to improve the limiting, local deformation prior to fracture in high strength materials, even though it be of less importance in our more ductile alloys.

In conclusion I should like to acknowledge indebtedness for assistance without which I could not have prepared this lecture. First to Carnegie Institute of Technology, and to Dr. Robert F. Mehl who as head of the Metals Research Laboratory there employed Frank Garofalo for the specific purpose of performing the additional work on the properties of steels that I have here re-

ported. My indebtedness to Dr. Mehl extends far beyond this. Mr. Garofalo gave me some help after we moved to the Pennsylvania State College. Mrs. Elizabeth Pearsall Hartner has helped me by reanalyzing the data and supervising the preparation of the illustrations; Mrs. Hartner's contributions to the lecture are many and to me invaluable. I have made free use of ideas concerning the analysis of tensile data that have been developed by Dr. John R. Low, Chief of the Division of Metallurgy at the Pennsylvania State College; Dr. Low and I have been associated for many years and it is impossible to allocate to him credit for all his specific contributions. The others who have helped must, because of their number, remain anonymous.

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PARTITION OF MOLYBDENUM IN HYPOEUTECTOID IRON-CARBON-MOLYBDENUM ALLOYS

By Fred E. Bowman

Abstract

The study of the partition of molybdenum between the ferrite and carbide resulting from the isothermal transformation of austenite as well as from tempering martensite is extended to hypoeutectoid alloys of iron, carbon, and molybdenum.

In these alloys the high molybdenum concentration in the carbides formed isothermally and subcritically from austenite proves that molybdenum must diffuse during the eutectoid reaction. This diffusion of molybdenum and the formation of the complex carbide $(Fe,Mo)_{23}C_6$ in alloys containing more than 0.50 per cent molybdenum provide an explanation for the retardation in the eutectoid reaction.

Partition of molybdenum is not, however, an explanation for the delay in formation of proeutectoid ferrite which molybdenum causes. The molybdenum concentration of the proeutectoid ferrite does not differ from that in the austenite and thus no diffusion of molybdenum is necessary in the transformation of austenite to ferrite. The importance of the proeutectoid reaction to hardenability is recognized and some explanations for the effect of molybdenum upon it are presented.

THE decomposition of austenite results in the formation of two phases: Ferrite containing approximately 0.01 per cent carbon, and carbide containing, theoretically, 6.67 per cent carbon. It is obvious that the diffusion of carbon is directly involved in the process. To a large extent then, the rate of isothermal transformation of austenite in alloy-free iron-carbon alloys must be influenced by the rate of diffusion of carbon in austenite. Molybdenum decreases the rate of transformation of austenite to pearlite at the higher subcritical temperatures, but to an extent much greater than can be ascribed to its influence upon the rate of diffusion of carbon. Ham, Parke, and

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Herzig (1)[1] have reported that in a eutectoid alloy at 1200 degrees Fahr. (650 degrees Cent.) 0.80 per cent molybdenum decreases the rate of carbon diffusion to about one-third that in unalloyed austenite, yet the same alloy content increases the time required for complete transformation at the same temperature by some 400 times (9). There is no doubt of the need for seeking the explanation of the effect of molybdenum upon the rate of transformation of austenite in some other phenomenon.

After careful consideration of various alternatives, an explanation based on the diffusion of molybdenum was selected as the most likely, since the rate of diffusion of molybdenum in austenite is much less than that of carbon. Calculations made available by J. L. Ham from his data on the rates of diffusion of carbon (1) and molybdenum (2) in austenite indicate that the ratio of the D value of carbon to that of molybdenum at 1200 degrees Fahr. (649 degrees Cent.) is roughly 500,000. If the transformation process requires the diffusion of the alloying element, the difference in order of magnitude of the D values of molybdenum and carbon strongly suggests that the rate of eutectoid austenite transformation depends upon the rate of diffusion of molybdenum.

It was reasoned that if, after transformation, the molybdenum concentration was shown to be higher in either the ferrite or carbide than in the parent austenite the diffusion of molybdenum would be established. Since molybdenum is a carbide-forming element, it was expected that any segregation of molybdenum would be in the carbide phase. Work with iron-carbon-molybdenum alloys of approximately eutectoid carbon content (3), (4) proved that this expectation was warranted. The molybdenum concentration was found to be much greater in the carbides than in the austenite from which these carbides formed, particularly in the alloys isothermally transformed at the higher subcritical temperatures. It appears that molybdenum dissolved in austenite so alters the free energy relationships between austenite, ferrite, and carbide at subcritical temperatures that its presence in the carbide is essential to the initiation and continuation of the isothermal transformation of eutectoid austenite. With the necessity for the presence of molybdenum in the carbide established, it became apparent that the rate of eutectoid austenite transformation, in the temperature range where the carbide is the nucleating phase, is dependent upon the rate of molybdenum diffusion.

[1]The figures appearing in parentheses pertain to the references appended to this paper.

Additional factors influence the rate of carbide formation (3), (4). In alloys containing more than 0.50 per cent molybdenum, cementite (Fe_3C) is not formed at the higher transformation temperatures. Instead, a face-centered cubic iron-molybdenum carbide, $(Fe,Mo)_{23}C_6$, having 116 atoms per unit cell is formed. Although this carbide had not previously been reported in connection with low molybdenum steels, it had been found and studied in some tool steel alloys (5). These studies revealed that the molybdenum atoms tend to occupy specific lattice sites. This tendency and the large number of atoms per unit cell suggest that the probability of formation of a $(Fe,Mo)_{23}C_6$ nucleus is much less than that of the smaller and less complex Fe_3C. Thus, not only does the rate of diffusion of molybdenum retard austenite transformation through a decreased rate of growth, but the very presence of molybdenum decreases the probability, and hence the rate, of nucleus formation.

After this description of the effect of molybdenum upon the eutectoid reaction in iron-carbon-molybdenum alloys was established, the investigation was extended to alloys of lower carbon content, wherein the formation of proeutectoid ferrite presents the possibility of another type of partition. The delay of the proeutectoid reaction is essential to complete hardening in hypoeutectoid alloys; therefore, the effect of molybdenum upon this reaction is as important as its effect upon the eutectoid reaction, insofar as hardenability is concerned. The results of an investigation of the partition of molybdenum in hypoeutectoid iron-carbon-molybdenum alloys and its relation to hardenability are the subject of this report.

Experimental Procedure and Results

The selection of the alloy compositions to be investigated was determined by the need for sufficient amounts of both the proeutectoid ferrite and the carbide for study. It was decided that alloys containing 0.40 per cent carbon were the most representative of the hypoeutectoid compositions that fulfilled these requirements. The molybdenum contents selected were the same as those used in the study of the eutectoid alloys.

As in the study of the eutectoid compositions, pure iron-carbon-molybdenum alloys were employed for the present work. They were prepared from electrolytic iron, pure graphite, and molybdenum powder by melting 1200-gram charges in zirconia crucibles and cast-

Table I
Analysis of the Electrolytic Iron Used in Preparing the Alloys

	Per Cent
Carbon	0.005
Manganese	0.002
Silicon	0.003
Phosphorus	0.001
Sulphur	0.004
Nickel	0.008
Molybdenum	<0.001
Lead	<0.001

ing in vacuo. The analysis of the electrolytic iron is given in Table I. Deoxidation was accomplished with carbon only. The metal was cast into a thick-walled steel mold at pressures of less than 0.01 millimeter of mercury. The ingots were forged to 1-inch rounds and machined to 7/8-inch o.d. to remove scale. The alloys were then homogenized at 2200 degrees Fahr. (1200 degrees Cent.) for 100 hours to eliminate any molybdenum concentration gradients. In the work with the eutectoid alloys such treatment had resulted in a satisfactory distribution of both the carbon and the molybdenum. In addition to the four molybdenum alloys prepared for the investigation, a molybdenum-free 0.40 per cent carbon alloy was prepared and treated in the same manner. This alloy was a standard for all subsequent work. The analyses of the ingots are listed in Table II.

Table II
Analysis of the Alloys

Alloy No.	Per Cent Mo	Per Cent C	Per Cent Si
1	<0.001	0.40	0.015
2	0.24	0.41	0.012
3	0.48	0.38	0.021
4	0.68	0.41	0.033
5	0.93	0.41	0.008

A sample of the iron-carbon alloy (Alloy No. 1) was completely transformed isothermally at 1300 degrees Fahr. (705 degrees Cent.) and furnace-cooled to insure complete carbon precipitation from ferrite. The lattice constant of the ferrite was carefully determined and the result ($a_o = 2.86063$ Å) indicated that the presence of silicon and other minor impurities did not affect the value for pure iron determined earlier (3).

Samples were cut from the homogenized bars for determination of the times required for the start and completion of both the pro-

eutectoid reaction and the carbide precipitation at constant subcritical temperature. The molybdenum content of the proeutectoid ferrite was determined by the X-ray diffraction method. The ferrite lattice parameters were calculated from X-ray diffraction patterns obtained in a 226.2-millimeter diameter back-reflection focusing camera. Unfiltered cobalt radiation was employed. After heat treatment the samples were surface ground and deeply etched in dilute nitric acid to remove worked metal. The gold powder used as a standard was held on the surface of the specimen by a thin film of oil. The gold diffraction pattern superimposed upon the ferrite pattern provided a means of adapting Cohen's method of calculation (6) to the determination of the ferrite lattice parameter. The molybdenum concentration in the ferrite was derived from the ferrite lattice parameter by means of the relationship between the two which was established and reported earlier (4).

The X-ray method requires that the samples be quenched from the austenite transformation temperature before the precipitation of any carbide. The quench results in a supersaturation of the ferrite with respect to carbon. The trapped carbon must be removed to prevent its effect upon the ferrite lattice parameter from masking that of the molybdenum. Consequently, a heat treatment was selected that would release this carbon from the ferrite yet leave the products of decomposition of martensite in a form that would not produce interference lines on the diffraction pattern. It was found, by experiment on a sample of the iron-carbon alloy (Alloy No. 1) partially transformed at 1300 degrees Fahr. (704 degrees Cent.) and water-quenched, that a treatment of ½ hour at 1000 degrees Fahr. (540 degrees Cent.) followed by a slow cool resulted in a ferrite with the lattice parameter of pure iron and no diffraction evidence of martensite decomposition products.

After these preliminaries, samples from each of the homogenized alloy bars were austenitized at 1600 degrees Fahr. (870 degrees Cent.) for ½ hour in deoxidized lead pots. From the austenitizing bath they were quenched directly into deoxidized lead pots at 1300 degrees Fahr. (705 degrees Cent.) and held for the length of time previously determined for producing as much proeutectoid ferrite as possible without precipitating any carbide. At the end of this time the transformation was abruptly halted by a water quench. The supersaturated condition of the ferrite, which resulted from the quench from 1300 degrees Fahr. (705 degrees Cent.), was relieved by treat-

ing the samples for ½ hour at 1000 degrees Fahr. (540 degrees Cent.) followed by a slow cool. Back-reflection X-ray patterns were made and the ferrite molybdenum contents determined from the lattice parameters. The results are given in Table III.

To establish the distribution which takes place during the precipitation of the carbide, samples of each alloy were completely transformed isothermally at 1300 degrees Fahr. (705 degrees Cent.). Post-transformation migration of molybdenum was prevented by quenching the samples into water immediately after the times established for complete transformation had elapsed. Excess carbon was removed from the ferrite by heating the samples to 1000 degrees Fahr. (540 degrees Cent.) for ½ hour followed by a slow cool. It was expected that the eutectoid reaction would yield a ferrite lower in molybdenum than that formed in the proeutectoid reaction and sufficiently homogeneous to provide a sharp diffraction pattern. If such a pattern could be obtained, it would be distinct from that of the proeutectoid ferrite, and a double ferrite line would be observed.

Table III
Distribution of Molybdenum Existing After Isothermal Transformation at 1300 Degrees Fahr.

Alloy No.	Composition Per Cent Mo	Composition Per Cent C	Transformation Times, Seconds Proeutectoid Start	Proeutectoid End	Complete	Per Cent Molybdenum in Proeutectoid Ferrite*	Per Cent Molybdenum in ** Carbide	Carbide Structure
2	0.24	0.41	15	60	480	0.24	0.75	Orthorhombic
3	0.48	0.38	35	360	800	0.47	1.78	Orthorhombic + Face-centered Cubic
4	0.68	0.41	50	630	3600	0.67	5.09	Face-centered Cubic
5	0.93	0.41	80	1000	7200	0.93	7.07	Face-centered Cubic

*Determined by X-ray diffraction.
**Determined by chemical analysis of electrolytically separated carbides.

Actually, however, no such distinction was obtained. The single ferrite line was extremely broad and no peaks of intensity could be detected. Calculations made from measurements taken from opposite sides of the line indicated that the molybdenum content of the total ferrite varied from that of the proeutectoid ferrite to zero. Metallographic examination revealed that, contrary to expectation, very little pearlite had been formed in any of the molybdenum alloys and that massive carbides had formed at the grain boundaries. The amount of pearlite decreased with increasing molybdenum content.

To provide samples of carbide for chemical analysis, homogenized bars of the iron-carbon-molybdenum alloys were isothermally transformed in a manner identical with that described for the X-ray samples. From these samples, the carbides were separated electrolytically in a 5 per cent solution of hydrochloric acid, using a current density of approximately 0.1 ampere per square inch. After washing in water, alcohol, and ether, part of the separated carbides were prepared for the powder diffraction camera and the remainder analyzed for molybdenum. The crystal structure of the carbides was determined from the diffraction pattern of each sample. The results are given in Table III.

The partition of molybdenum in each alloy was investigated in an identical manner after transformation at 1200 degrees Fahr. (650 degrees Cent.). Here again it was impossible to obtain any information concerning the molybdenum content of the eutectoid ferrite by X-ray diffraction methods. Although the massive carbides appeared after isothermal transformation at this temperature, the amount of pearlite had increased. Carbides were separated and analyzed chemically. Results of the study of the transformation at 1200 degrees Fahr. (650 degrees Cent.) are given in Table IV.

Table IV
Distribution of Molybdenum Existing After Isothermal Transformation at 1200 Degrees Fahr.

Alloy No.	Composition Per Cent Mo	Composition Per Cent C	Transformation Times, Seconds Proeutectoid Start	Transformation Times, Seconds Proeutectoid End	Transformation Times, Seconds Complete	Per Cent Molybdenum in Proeutectoid Ferrite*	Per Cent Molybdenum in ** Carbide	Carbide Structure
2	0.24	0.41	5	35	400	0.23	1.19	Orthorhombic
3	0.48	0.38	15	400	850	0.48	1.98	Orthorhombic + Face-centered Cubic
4	0.68	0.41	300	820	2400	0.67	4.10	Face-centered Cubic
5	0.93	0.41	420	2400	6000	0.93	6.25	Face-centered Cubic

*Determined by X-ray diffraction.
**Determined by chemical analysis of electrolytically separated carbides.

Microscopic investigation of the alloys transformed at 1100 degrees Fahr. (595 degrees Cent.) revealed an extremely fine pearlitic structure with no appreciable amount of proeutectoid ferrite. The structures were similar to those found in the eutectoid alloys transformed at 1000 degrees Fahr. (540 degrees Cent.) and no X-ray ex-

amination could be made. Apparently both the ferrite and carbide were so finely dispersed that no satisfactory diffraction pattern was obtainable. The times required to complete the transformation of these alloys at 1100 degrees Fahr. (595 degrees Cent.) indicate that this is the region of most rapid transformation.

The partition of molybdenum resulting from the tempering of martensite is of both theoretical and practical interest. This treatment affords an opportunity to study the approach to equilibrium from a direction different from that provided by isothermal transformation, and the rate at which molybdenum enters the carbide from ferrite is of importance in the elevated temperature applications of molybdenum steels. In order to obtain data comparable to those secured from the eutectoid alloys, samples of each homogenized alloy were quenched from 1600 degrees Fahr. (870 degrees Cent.) into water after ½ hour at temperature. These quenched samples were then held at 1300 degrees Fahr. (705 degrees Cent.) for 150 hours in vacuo. The pressure was maintained at from 0.003 to 0.006 millimeters of mercury to avoid surface decarburization. Since the alloys were essentially oxide-free, there was little or no possibility of internal decarburization by the reduction of oxide inclusions. After the lattice parameter of the ferrite was determined in the back-reflection camera, the samples were replaced in the furnace for an additional 100 hours. The ferrite lattice was again measured after this treatment. In addition, after each tempering period, the crystal structure of the carbides was determined in the powder camera directly from the sample. The results are listed in Table V.

Attempts were made to carry out similar investigations after tempering at 1200 degrees Fahr. (650 degrees Cent.) and 1100 degrees Fahr. (595 degrees Cent.); however, as in the eutectoid alloys, partial graphitization was encountered in all but the highest molybdenum alloy after 150 hours. The results obtained from Alloy No. 5, tempered at 1200 degrees Fahr. (650 degrees Cent.), are shown in Table VI.

Discussion of the Results

The time required for the start of the proeutectoid reaction at 1300 degrees Fahr. (705 degrees Cent.) in the 0.94 per cent molybdenum alloy is five times that in the 0.24 per cent molybdenum alloy, but the molybdenum concentration in the ferrite in each case is es-

sentially the same as in the austenite from which it formed. Therefore, the delay caused by molybdenum in the formation of proeutectoid ferrite cannot be attributed to the diffusion of molybdenum, and the explanation must be sought in some less direct influence of the alloying element. There are two such effects of molybdenum that can be proposed at this time.

Since, in the hypoeutectoid alloys studied, a ferrite containing approximately 0.01 per cent carbon forms directly from austenite containing 0.40 per cent carbon, carbon must diffuse during the process. Ham, Parke and Herzig (1) have demonstrated that the rate of diffusion of carbon in austenite below 2000 degrees Fahr. (1095 degrees Cent.) is decreased by molybdenum. That the amount of decrease is slight may be illustrated by calculations made from the data of Wells and Mehl (7) for a molybdenum-free, 0.44 per cent carbon alloy and from the data of Ham, Parke and Herzig for an 0.80 per cent molybdenum alloy of similar carbon content. The results predict that at 1300 degrees Fahr. (705 degrees Cent.) the diffusion coefficient for carbon in the molybdenum-bearing austenite is only two-thirds that in the unalloyed austenite. A comparison of this value for the reduction in diffusion coefficient induced by molybdenum and the 80 per cent reduction in rate of proeutectoid formation produced indicates little, if any, connection between the two. However, the rate of transfer of matter by diffusion is not determined by diffusion coefficients alone. The rate of diffusion is also influenced by concentration gradients and the manner in which molybdenum affects the carbon concentration gradient at the austenite-ferrite interface may be the controlling factor in the rate of ferrite formation.

No attempt has as yet been made to establish quantitatively the extent to which molybdenum affects the gamma-to-alpha transformation in carbon-free iron-molybdenum alloys. It is known that nickel and manganese render the allotropic transformation in iron extremely sluggish, and Ham (2) has suggested that their retardation of the rate of austenite transformation is the result of this effect. The case for molybdenum is admittedly different. An inspection of the appropriate phase diagrams shows that molybdenum forms a gamma loop with iron and hence would seem to stabilize ferrite. Nickel and manganese, on the other hand, render austenite stable at temperatures where ferrite is the stable phase in pure iron. However, Austin and Pierce (8) report an appreciable time factor in the gamma-to-alpha transformation in iron-chromium alloys. Like that of iron and mo-

lybdenum, the iron-chromium diagram displays a gamma loop. Molybdenum then may also affect the rate of formation of proeutectoid ferrite through its effect upon the gamma-to-alpha transformation.

In contrast to that of the eutectoid reaction, little attention has been given to the nucleation rate of the proeutectoid reaction. Molybdenum may act to retard this rate in several ways. If it should be shown that the solubility of carbon in ferrite is decreased by molybdenum, the time required to form a nucleus of ferrite sufficiently low in carbon would be longer than for a similar nucleus of ferrite in a molybdenum-free alloy. The rate of growth of ferrite requiring a lower carbon concentration would also be reduced, since more carbon would have to be transported.

There are, then, at least two possible ways by which molybdenum can retard the proeutectoid reaction, acting singly or in combination to reduce either or both the rate of nucleation and rate of growth: (a) An increase in the time necessary to establish the correct ferrite carbon concentration through the influence of molybdenum upon the D value for carbon in austenite, through an effect upon the carbon concentration gradients at the austenite-ferrite interface and/or a decrease in the solubility of carbon in ferrite, and (b) an interference in the rearrangement of iron atoms required by the change from the face-centered to the body-centered cubic crystal lattice and thus a retardation in the allotropic transformation rate.

The validity of these proposals will be tested in future experimental work.

The extent of molybdenum concentration in the carbide phase of the hypoeutectoid alloys (Tables III and IV) is consistent with the results of the investigation of the eutectoid alloys, and demonstrates the diffusion of the alloying element during the eutectoid transformation. As was the case in the eutectoid carbon alloys, the face-centered cubic, $(Fe,Mo)_{23}C_6$, is formed in hypoeutectoid alloys containing more than 0.50 per cent molybdenum.

No measurement of the molybdenum contents of the eutectoid ferrites could be made. It was noted earlier that the carbides occur in a massive state, in contrast to the lamellar pearlites in steels of commercial analyses and also in the eutectoid alloys previously studied (4). The manner in which these carbides form may be the result of a greatly decreased rate of nucleation. It has already been proposed that the presence of molybdenum and the necessity for its appearance in the carbide is sufficient reason for expecting a decrease

in the probability of formation of a suitable carbide nucleus. The lower carbon content would produce a further decrease in rate of nucleation from that for alloys of higher carbon content. The combination of these two factors apparently decreases the rate of nucleation in the alloys under investigation to the point where intermittent formation of carbide nuclei required to form pearlite is more difficult than continued growth from a single carbide nucleus. The carbon concentration gradients set up in the austenite through the rejection of carbon by proeutectoid ferrite would also tend to change the mode of transformation from that in a homogeneous austenite. The eutectoid reaction is further complicated in that a relatively large amount of free ferrite is formed after carbide precipitation has begun. The results of X-ray diffraction work on the completely transformed alloys indicate that the partition of molybdenum between ferrite and carbide in the hypoeutectoid alloys is by no means as distinct as in the eutectoid alloys. This is not surprising in view of the obvious difference in the mode of formation of the carbides. The area of contact between the austenite, ferrite, and carbide is much less in the case of the massive type of carbide formation. As a result there is less opportunity for equilibrium to be established with respect to concentration gradients.

It is notable that the molybdenum contents of the carbides formed at 1200 degrees Fahr. (650 degrees Cent.) in Alloys 2 and 3 are greater than the molybdenum contents of those formed at 1300 degrees Fahr. (705 degrees Cent.). It is believed that the difference in molybdenum contents can be accounted for by the difference in the manner in which the carbide precipitates from the austenite. It was pointed out that after transformation at 1300 degrees Fahr. (705 degrees Cent.) the carbides occur more or less massively at the grain boundaries. After transformation at 1200 degrees Fahr. (650 degrees Cent.) the carbides are more finely dispersed and the tendency for the formation of lamellar pearlite is greatly increased. The greater surface area of these carbides must afford more opportunity for the transfer of molybdenum atoms from the austenite to the carbides as the latter are forming.

In general the results of the investigation of tempered martensite in the hypoeutectoid alloys substantiate the findings reported from the work with eutectoid alloys (3), (4). The molybdenum concentration of the carbides is, in the hypoeutectoid alloys, higher than in the eutectoid materials, but this is not surprising, for a more rapid

rate of transfer from the ferrite to carbide would be expected in alloys having a higher molybdenum: carbon ratio, solely on the basis of concentration gradients. This rate of transfer is greater in the early stages of tempering treatment, as evidenced by the increase in molybdenum found in carbides after 150 hours as compared with that found after 250 hours. If it is assumed that the molybdenum concentration in the carbides, as they originally form from martensite, equals that of the martensite, the increase of molybdenum in the carbide in the first 150 hours is a matter of 400 to 500 per cent, while in the next 100 hours it is approximately 35 per cent.

One result of the tempering experiments deserves special attention. It will be noted from Table V that in the two higher molybdenum alloys (Alloys 4 and 5) the molybdenum content of the carbides

Table V

Alloy Quenched from 1600 Degrees Fahr. and Tempered at 1300 Degrees Fahr.

Alloy No.	Per Cent Molybdenum in Ferrite		Per Cent Molybdenum in Carbide		Carbide Structures	
	After 150 Hr.	After 250 Hr.	After 150 Hr.	After 250 Hr.	After 150 Hours	After 250 Hours
2	0.17	0.15	1.15	1.50	Orthorhombic	Orthorhombic
3	0.34	0.31	2.31	2.81	Orthorhombic	Orthorhombic
4	0.48	0.44	3.43	4.01	90% Orthorhombic 10% Face-centered cubic	90% Orthorhombic 10% Face-centered cubic
5	0.62	0.58	4.79	5.33	80% Orthorhombic 20% Face-centered cubic	80% Orthorhombic 20% Face-centered cubic

is appreciably above 3 per cent. The results of work on the eutectoid alloys led to the assumption that the transition of carbide structure from orthorhombic to face-centered cubic occurred at a molybdenum concentration of from 3 to 3½ per cent. No orthorhombic carbides containing more than 3 per cent molybdenum were found in the eutectoid alloys, irrespective of whether they were formed directly from austenite isothermally or by tempering martensite. In the high molybdenum hypoeutectoid alloys (Alloy No. 5, Table V) on tempering martensite at 1300 degrees Fahr. (705 de-

Table VI

Alloy Containing 0.93 Per Cent Molybdenum, Quenched from 1600 Degrees Fahr. and Tempered for 150 Hours at 1200 Degrees Fahr.

Per Cent Mo in Ferrite	Per Cent Mo in Carbide	Carbide Structure
0.74	2.81	Orthorhombic

grees Cent.) a carbide forms that is predominantly orthorhombic, yet contains 5.33 per cent molybdenum. Obviously, by tempering martensite at 1300 degrees Fahr. (705 degrees Cent.) the molybdenum content of the orthorhombic carbide can be raised above the value which, on isothermal transformation of the alloy at this temperature, results in the face-centered cubic carbide.

Partition and Hardenability

The investigation of the partition of molybdenum between the ferrite and carbide was prompted by the need for explanation of the effect of molybdenum upon the rate of transformation of austenite. Since this effect determines the value of molybdenum in increasing the hardenability of steel, an understanding of the role of the alloying element in the transformation process is essential to its most efficient use.

From S-curve data it is clearly evident that the transformation of austenite to the carbide-ferrite aggregate is retarded by molybdenum (9), (10). The molybdenum concentration in the carbides formed by isothermal transformation of molybdenum-bearing austenite is much greater than that in the parent austenite. This segregation of molybdenum in the carbide proves that the diffusion of the alloying element is involved in the transformation. It has been shown that the rate of this transformation of austenite to pearlite, then, must be influenced by the rate of diffusion of molybdenum.

In the temperature region where molybdenum exerts its greatest effect upon the rate of eutectoid transformation, the reaction is nucleated by carbide. The necessity for the presence of molybdenum in these nucleating carbides results in a retardation of nucleation rates to an extent governed by the rate at which molybdenum atoms can reach their required positions. Since nucleus formation is a matter of probability, the process is further delayed by the formation, in alloys containing more than 0.50 per cent molybdenum, of a large complex face-centered cubic iron-molybdenum carbide, $(Fe,Mo)_{23}C_6$, containing 116 atoms per unit cell. It is not difficult to visualize the decreased likelihood of formation of a group of 116 iron, carbon, and molybdenum atoms in the correct proportions and arrangement as compared with the 16 iron and carbon atoms required for the cementite unit cell formed in iron-carbon alloys.

Once formed, the nuclei grow, as pointed out by Ham (2), at a

rate determined by the rapidity with which molybdenum atoms reach and cross the austenite-carbide interface. Thus the rate of growth also is influenced by the rate of diffusion of molybdenum in austenite.

In hypoeutectoid alloys, the rate of formation of proeutectoid ferrite may be more important than the eutectoid reaction rate in determining the size of section that can be completely transformed to martensite. The proeutectoid reaction is retarded by the presence of molybdenum, although to a lesser extent than is the eutectoid reaction. The present work has shown that the decreased rate of the proeutectoid reaction cannot be accounted for on the basis of partition, for the molybdenum concentration in the proeutectoid ferrite remains essentially the same as in the austenite from which it forms. There is still need, then, for an explanation of the part played by molybdenum in this reaction; the explanations discussed previously will be investigated. Knowing how molybdenum affects the austenite-to-ferrite transformation becomes increasingly important if it is borne in mind that the upper bainite reaction, which occurs in the region of most rapid transformation, and hence actually determines the extent to which full hardening can be attained, is nucleated by ferrite.

Summary

The investigation of the partition of molybdenum in iron-carbon-molybdenum alloys has established the following facts:

(a) Molybdenum segregates in the carbide during the isothermal transformation of austenite at the higher subcritical temperatures.

(b) In alloys containing more than 0.50 per cent molybdenum and isothermally transformed at 1200 and 1300 degrees Fahr. (650 and 705 degrees Cent.) the cementite is replaced by an iron-molybdenum carbide, $(Fe,Mo)_{23}C_6$, having a larger and more complex unit cell.

(c) Molybdenum retards the rate of formation of proeutectoid ferrite, although the molybdenum content of the ferrite remains unchanged from that of the austenite.

(d) Molybdenum moves from the ferrite to the carbide during the tempering of martensite.

(e) The rate of graphitization of the carbides during the tempering of martensite is retarded by molybdenum.

(f) The molybdenum concentration in the orthorhombic carbide (Fe_3C) can be increased by tempering martensite to a value greater than that which results in the presence of the face-centered cubic carbide, $(Fe,Mo)_{23}C_6$, after the isothermal transformation of austenite.

The segregation of molybdenum in the carbides formed by the isothermal transformation of austenite at 1300 and 1200 degrees Fahr. (705 and 650 degrees Cent.) demonstrates the diffusion of molybdenum during the eutectoid transformation process. The delay effected by molybdenum in the eutectoid reaction is accounted for by the slow rate of diffusion of molybdenum as compared with that of carbon, which determines the rate of transformation in unalloyed austenite. In addition, under conditions favoring a relatively high molybdenum concentration in the carbide, the rate of carbide nucleation and growth is further decreased by the formation of a large and complex face-centered cubic carbide of iron and molybdenum, $(Fe,Mo)_{23}C_6$. There is consequently now established an explanation for the effect of molybdenum upon the hardenability of steel insofar as the hardenability is determined by the eutectoid reaction.

The present work has demonstrated the need for further investigation of the influence of molybdenum upon the austenite-to-ferrite transformation. The diffusion of molybdenum is clearly not the explanation, for no change in molybdenum concentration occurs as proeutectoid ferrite forms isothermally from austenite. The effect of molybdenum upon the solubility of carbon in ferrite, upon the rate of gamma-to-alpha transformation in carbon-free iron-molybdenum alloys, and upon the rate of transport of carbon are presented as possible explanations.

Acknowledgment

The author wishes to acknowledge his indebtedness to the members of the staff of the Climax Molybdenum Co. research laboratory, who have contributed much to the partition investigation through the preparation of alloys and through their advice and suggestions during the experimental work and preparation of the paper.

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DISCUSSION

Written Discussion: By F. R. Morral, Metal Trades Laboratory, Technical Service and Development Div., American Cyanamid Co., Stamford, Conn.

The series of papers on iron-molybdenum alloys Mr. Bowman has been presenting in the last 2 to 3 years has been of special interest to the writer.

In a previous discussion it was pointed out[2] that molybdenum or tungsten in small amounts was necessary to secure the formation of the $(Fe,Mo)_{23}C_6$ and $(Fe,W)_{23}C_6$ face-centered cubic compounds. This work indicates that as little as 0.48 per cent molybdenum and 0.38 per cent carbon is sufficient to cause the presence of the above carbide. We found some 12 years ago that tungsten double carbide was found with 0.76 per cent tungsten and 0.36 per cent carbon. Less amounts may be sufficient. It is possible that this carbide forms in preference to the orthorhombic cementite because of the similarity of the lattices between the double carbide and austenite. It may be interesting to note that carbides similar in structure are present in the Cr-Fe-C and Mn-Fe-C systems. In the first it was believed that 25 per cent may be iron,[3] while in the

[2]F. R. Morral, Discussion following article "The Alpha Iron Lattice Parameter as Affected by Molybdenum, and an Introduction to the Problem of the Partition of Molybdenum in Steel," by F. E. Bowman, R. M. Parke and A. J. Herzig, TRANSACTIONS, American Society for Metals, Vol. 31, 1943, p. 496.

[3]A. Westgren, G. Phragmen and T. R. Negresco, "On the Structure of the Iron-Chromium-Carbon System," *Journal*, Iron and Steel Institute (London), Vol. 117, No. 1, 1928, p. 383.

latter up to 50 per cent has been found.[4] In all cases, the carbon present is slightly less than 6 per cent. In these systems iron only substitutes a few of the atomic places corresponding to, for example, the chromium in the $Cr_{23}C_6$ lattice.[5]

Mr. Bowman has investigated what happens on isothermal transformation at various temperatures and on tempering. The above $(Fe,Mo)_{23}C_6$ seems to form in all cases. He has been interested in determining what the maximum and minimum values for molybdenum may be in this carbide and cementite. I wonder if treatment of the carbide separated electrolytically and heated in vacuum may not give an answer. On experiments of this nature the writer made some years ago on the $(Fe,W)_{23}C_6$ he found it to decompose into Fe_3C and WC. This technique may also give a clue as to the maximum Mo $(Fe,Mo)_{23}C_6$ and Fe_3C may hold in solid solution.

Kelley[6] has computed the enthropy of $Cr_{23}C_6$. This compound was shown to be stable at temperatures below its melting point. A technique such as used by Kelley[7] to determine the partition of carbon between iron and manganese in the γ range may be applicable to the work being done by the author on the Fe-Mo-C system. Has Mr. Bowman followed the change in the lattice parameter of $(Fe,Mo)_{23}C_6$ for his various alloys? He has reported values for molybdenum up to 7.07 per cent by weight. By X-ray diffraction it might be possible to know how much molybdenum is present in the face-centered cubic carbide in a certain alloy after a given treatment. These data I believe, because of their fundamental nature, would be valuable in future work. Mr. Bowman is to be congratulated for presenting a very interesting piece of work.

Oral Discussion

E. O. KIRKENDALL:[8] The author has picked a problem of very great practical importance. Little is known of how alloying elements produce an increase in hardenability of steel. Mr. Bowman is to be highly complimented on his ability to break the complex problem up into steps and handle each part vigorously.

This paper adds valuable information regarding the mechanism of the increase in hardenability of steel due to molybdenum and, without a doubt, to the general problem of the mechanism of increasing the hardenability of steel. Mr. Bowman has shown that molybdenum does not have to diffuse to permit the precipitation of proeutectoid ferrite, nor that complicated nuclei are required to form proeutectoid ferrite. Molybdenum, therefore, does not retard the precipitation of proeutectoid ferrite as much as it retards the carbide nucleated pearlite reaction. Other than the chemical attraction between molyb-

[4]E. Ohman, "The Carbides in the Iron-Manganese-Carbon System," *Jernkontorets Annaler*, Vol. 128, 1944, p. 13-16.

[5]A. Westgren, "The Crystal Structure of High Speed Steel Carbide," *Jernkontorets Annaler*, Vol. 117, 1933, p. 501-512.

[6]K. K. Kelley, F. S. Boericke, G. E. Moore, E. H. Huffman and W. M. Bangert, "Thermodynamic Properties of Carbides of Chromium," U. S. Bureau of Mines, Technical Paper 662, 1944.

[7]K. K. Kelley, "Data on Theoretical Metallurgy. Part VIII, Thermodynamic Properties of Metal Carbides and Nitrides," U. S. Bureau of Mines, Bulletin 407, 1937.

[8]College of Engineering, Wayne University, Detroit. Then assistant professor, now secretary, Metals Divisions, American Institute of Mining and Metallurgical Engineers.

denum and carbon, is there any reason why molybdenum is not forced to segregate in the ferrite to permit the simple orthorhombic cementite precipitation and therefore a more rapid pearlite reaction than observed? In this connection, it might be interesting to obtain data on the effect of molybdenum on the diffusion of carbon from a high molybdenum-high carbon-iron alloy into pure iron, or from a molybdenum-free high carbon-iron alloy into a carbon-free high molybdenum alloy.

J. H. HOLLOMON:[9] I should like to ask Mr. Bowman a question which not only arises from the discussion of this paper but from previous papers on the partition of molybdenum between the various constituents of the composition of austenite. Fundamentally, the suggestion had been made that since the carbide phase of the ferrite-carbide aggregate is higher in molybdenum content than the carbide phase in nonmolybdenum-bearing steels, it is, therefore, necessary in the formation of a ferrite carbide aggregate for the molybdenum to diffuse from the austenite of lower molybdenum concentration to carbide of higher molybdenum concentration. The question still exists in my own mind, why is it necessary for the molybdenum to diffuse to the carbide to form a high molybdenum content carbide? Even if the molybdenum carbide is more stable, thermodynamically, there is no reason why Fe_3C cannot form. Certainly the probability of forming Fe_3C is greater than the probability of forming any complex molybdenum carbide such as $(Fe,Mo)_{23}C_6$. If molybdenum carbide must form then of course molybdenum must diffuse to form the carbide, but Fe_3C can form from an austenite of high molybdenum content and I should like to ask the question why it does not. For if Fe_3C forms, then obviously molybdenum would not have to diffuse and the decomposition of the austenite could occur just as rapidly as in plain carbon steels. We know that Fe_3C can form below the critical temperature, for if a steel bearing molybdenum is quenched to martensite and tempered at low temperatures, the product of the tempering is Fe_3C and ferrite. Now, if the same argument was to hold, then we must say that there is some reason why at low tempering temperatures Fe_3C forms and at high tempering temperatures a molybdenum carbide forms, but why does not a molybdenum carbide form at low temperatures? In other words, it seems to me, that it is still necessary to explain why Fe_3C does not form in high molybdenum alloys. The mere statement that there is a high concentration of molybdenum in the carbide phase does not explain the retarding effect of molybdenum on the pearlite transformation. The full explanation must involve an understanding of the reasons why Fe_3C does not occur.

W. A. PENNINGTON:[10] In studies involving the decarburization of steels, I have done a little experimenting with different chemical elements, one of which was molybdenum. While the investigations have not been extensive, they have been broad enough to throw some light upon the effect of molybdenum upon the thermodynamic equilibrium end point.

With molybdenum it was found that exactly the same end point concentration of carbon was obtained as in the case of alloy-free steel. This behavior is entirely different from that of titanium and vanadium. These elements give end

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[10]Chief chemist and metallurgist, Carrier Corporation, Syracuse, N. Y.

points which depend upon the relative amount present. I do not know what this means, but it does suggest that titanium and vanadium occur as insoluble carbides and that molybdenum is in solution in the ferrite.

It is possible that the molybdenum occurs in the ferrite solution as dissolved molybdenum carbide which decomposes during the decarburizing heat treatment to satisfy the same thermodynamic conditions as though the molybdenum were not present. Then, on the other hand, it may occur as simple dissolved elementary molybdenum.

Author's Reply

The author wishes to express his appreciation to the discussers for their interest in the subject of partition.

The results of Dr. Pennington's decarburization studies are extremely interesting. The "end point concentration of carbon" is, I assume, the carbon content of the ferrite after partial decarburization of a sample at some temperature below or within the critical range. The fact that Dr. Pennington finds this value to be the same in molybdenum-bearing steels as in plain carbon steels indicates that the presence of molybdenum does not essentially alter the solubility of carbon in ferrite. Or, to put it another way, the presence of molybdenum in the carbides does not alter their solubility in ferrite, whereas the presence of titanium and vanadium apparently does.

In reply to Dr. Kirkendall's question, it seems that an attraction between the carbon and molybdenum atoms in austenite provides the most simple explanation for the presence of molybdenum in the carbides formed directly from austenite. If such an attraction exists, the necessity for the diffusion of molybdenum during the transformation of austenite is apparent. The molybdenum atoms would actually be dragged into position by the carbon atoms moving to form the carbide. It is planned to carry out diffusion experiments similar to those suggested by Dr. Kirkendall, to determine whether or not there is an attraction between molybdenum and carbon in austenite. If such is the case, the diffusion coefficient for carbon moving out of an iron-carbon-molybdenum alloy into pure iron would certainly be less than that determined from a diffusion couple in which the molybdenum contents were the same on either side of the interface. The latter condition has prevailed in all previous determinations of the effect of molybdenum upon the rate of diffusion of carbon in austenite.

Dr. Morral's suggestions have proved very valuable in the past, and certainly those contained in the present discussion are no exception. With regard to following the change in lattice parameter of $(Fe,Mo)_{23}C_6$ with varying molybdenum contents, it has been found that in the carbides of this type which have thus far been encountered no change detectable from the powder pattern has been produced. Any variation in the lattice dimensions must necessarily be determined from back-reflection patterns. The lines from the iron-molybdenum carbide which fall in this range are extremely weak, making precision measurements difficult.

Captain Hollomon is to be complimented upon a particularly valuable

discussion. The points he raises are certainly worthy of further thought and consideration.

Captain Hollomon's desire for an explanation of why molybdenum must diffuse to the carbide during the transformation of austenite is no greater than our own. Unfortunately, the data available at present are insufficient for us to formulate an explanation. However, inability to explain the phenomenon does not permit us to disregard the facts.

Intuitive reasoning is an invaluable tool; yet danger lies in the tendency to cling to the conclusions reached by such a process in the face of contradictory evidence. By intuition, Captain Hollomon concludes that the probability of forming cementite from austenite containing 0.50 per cent molybdenum is greater than the probability of forming the complex $(Fe,Mo)_{23}C_6$. Experimental evidence indicates that while direct transformation of an austenite containing 0.30 per cent molybdenum or less results in the formation of cementite, an increase to 0.50 per cent molybdenum results in the formation of the complex carbide. That which forms must be the most probable thing to form and we have adjusted our thinking accordingly.

Two conditions which could explain the results can be proposed at this time: There may be, as Dr. Kirkendall suggests, an attraction between the molybdenum and carbon atoms in austenite. Such an attraction could require the simultaneous arrival of carbon and molybdenum atoms at a carbide site. It is also possible that the energy barrier to carbon atoms at the austenite-cementite interface is increased by molybdenum to an extent that it becomes greater than that at the austenite-$(Fe,Mo)_{23}C_6$ interface. The magnitude of the difference in these barrier energies would determine the degree to which the formation of the complex carbide would be favored.

EFFECT OF VARIATIONS IN COMPOSITION AND HEAT TREATMENT ON SOME PROPERTIES OF 4 TO 6 PER CENT CHROMIUM STEEL CONTAINING MOLYBDENUM AND TITANIUM

By George F. Comstock

Abstract

The hardness, tensile properties, notch impact resistance, microstructure, weldability, high temperature oxidation, and time for rupture under stress at high temperature, are reported for 31 different 5 per cent chromium-molybdenum-titanium steels, some of commercial manufacture and some of special composition melted in the laboratory, heat treated in several different ways. The effects of variations in manganese, silicon, phosphorus, molybdenum, nitrogen, aluminum, and the titanium-carbon ratio are discussed. It is concluded that titanium-carbon ratios between 3.5 and 5.5 give better properties in general, and especially better notch toughness and high-temperature rupture strength than can be obtained with higher titanium, and such ratios are sufficient for acceptable restriction of air hardening. Silicon around 1 per cent is very beneficial to the oxidation resistance of this steel, and can be used in the steels with titanium-carbon ratio not over 5.5, without impairing other properties if the manganese content is also raised to about 0.75 or 1 per cent and the steel is merely tempered at about 1350 degrees Fahr. (730 degrees Cent.) without annealing or normalizing, after hot working at a reasonably low temperature to produce a fine grain size. Higher than normal phosphorus, molybdenum, aluminum, and nitrogen were found to be undesirable.

ALTHOUGH there is an adequate amount of published data (1), (2), (3), (4)[1] available on the high temperature properties of steel containing about 4 to 6 per cent chromium and 0.5 per cent molybdenum, which is quite widely used in the oil and power industries, similar data on this grade of steel containing titanium are

[1]The figures appearing in parentheses pertain to the references appended to this paper.

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Table I
Published Creep Data on 5 Per Cent Chromium-Molybdenum-Titanium Steel

Source of Data	Babcock and Wilcox (5)	Timken (6)	U. S. Steel (3), (7)
Stress for 0.1 Per Cent Creep per 1000 Hours at 1000 Degrees Fahr.	13000	12000	11500
Per Cent of Increase Over Steel Without Titanium	41	19	26
Stress for 0.1 Per Cent Creep per 1000 Hours at 1200 Degrees Fahr.	4700	1500	.....
Per Cent of Increase Over Steel Without Titanium	161	—46	.....
Stress for 0.01 Per Cent Creep per 1000 Hours at 1000 Degrees Fahr.	7000	7000	.....
Per Cent of Increase Over Steel Without Titanium	—3	—8	.....
Stress for 0.01 Per Cent Creep per 1000 Hours at 1200 Degrees Fahr.	1250	850	.....
Per Cent of Increase Over Steel Without Titanium	39	—50	.....

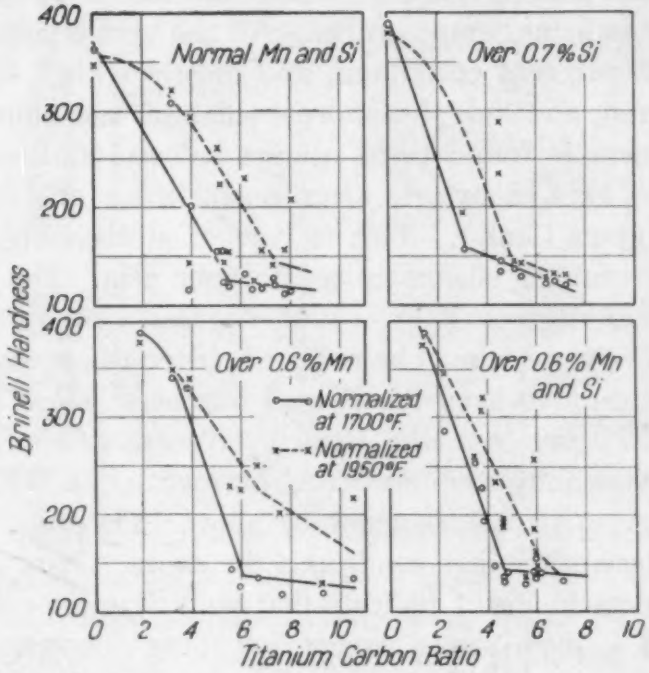
scarce. This is especially true of stress-rupture data, or measurements of the time required for breaking under different stresses and temperatures, although such data have been stated to be specially useful in the design of high temperature tubing. Creep data on this type of steel have been reported by a few investigators, but with some variations, as indicated in Table I.

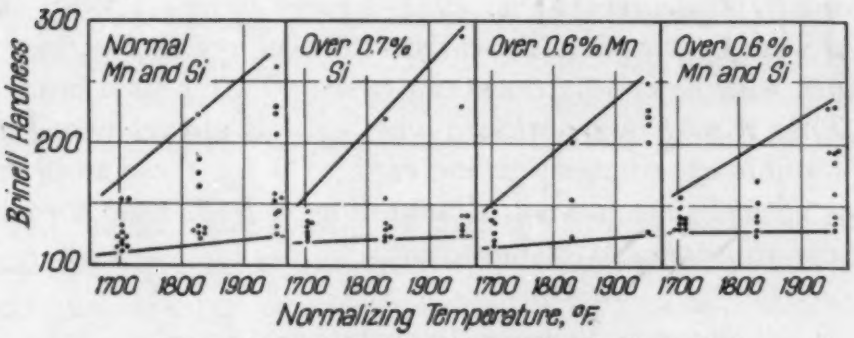
Stress-rupture tests on this steel have been reported, as far as the author can find, only by Dr. Clark and his collaborators at the Timken Steel Research Laboratory and the University of Michigan (1), (6). They indicate that the stress versus time-for-rupture curves of the titanium steel are below the corresponding curves for steel containing no titanium at all temperatures, and the stress for rupture in 100,000 hours is likewise lower for the titanium steel at all temperatures from 1000 to 1300 degrees Fahr. (540 to 705 degrees Cent.). These results were obtained with specimens annealed at 1600 degrees Fahr. (870 degrees Cent.) annealing being required for the nontitanium steel to prevent excessive hardness and low ductility. They do not seem to agree any too well with the published creep values, especially those determined for the Babcock & Wilcox Co. at Massachusetts Institute of Technology (5), and the investigation here reported was started to determine whether changes in the composition or heat treatment of the titanium steel might not improve the high temperature properties.

Results of Hardness Tests

The principal reason, if not the only one, for using titanium in this 5 per cent chromium, 0.5 per cent molybdenum steel is to prevent air hardening, or the occurrence of a hard martensitic structure

lacking ductility, after the steel is heated for such purposes as the fabrication or bending of tubes, welding, etc. This steel without titanium has to be annealed and cooled very slowly after such processing to retain machinability and avoid brittleness, while with the





(For Steels of Over 4 Ti/C Ratio Exclusively.)

Fig. 1—Brinell Hardness of Normalized 5 Per Cent Chromium-Molybdenum-Titanium Steels.

proper amount of titanium added these properties are obtained after cooling at normally rapid rates in air. The use of titanium for this purpose in chromium steels has already been reported and explained (8), (9), (10), but the effects of variations in composition and heat treatment on air hardening or its prevention may not be so well known. The results of our hardness tests are therefore reported as an introduction to the discussion of other properties.

Brinell hardness results on 40 experimental and 8 commercial heats of 5 per cent chromium-molybdenum-titanium steel, normalized at several temperatures, are reported graphically in Fig. 1. These data are separated according to the manganese and silicon contents of the steels, and plotted against the titanium-carbon ratios, and also against the normalizing temperatures. All the results apply to steels with 4.5 to 5.8 per cent chromium, and mostly with 0.4 to 0.6 per cent molybdenum, and low phosphorus, sulphur, and aluminum.

The hardness is found to be always below 160 Brinell with a titanium-carbon ratio above 4, after normalizing at 1700 degrees Fahr. (925 degrees Cent.). This is true of steels with manganese up to 1.49 per cent, or silicon up to 1.58 per cent. But after normalizing at 1950 degrees Fahr. (1065 degrees Cent.), a titanium-carbon ratio of 6 or more may be required, especially with manganese above 0.70 per cent, to keep the Brinell hardness below 200. With silicon above 0.70 per cent, the Brinell hardness of the 5 per cent chromium-molybdenum-titanium steel, normalized at 1700 degrees Fahr. (925 degrees Cent.), is generally above 125, while with lower silicon and otherwise similar conditions the average hardness is less. The hardness data in Fig. 1 indicate that with titanium-carbon ratios between 3 and 4, these steels, except for those abnormally high in manganese, are not generally harder than 250 Brinell after normalizing from 1700 degrees Fahr. (925 degrees Cent.). With silicon above 0.75 per cent, a titanium-carbon ratio as low as 3 seems to be permissible with Brinell hardness not over 160 after such heat treatment. Thus it may be questioned whether it is always necessary to specify a minimum titanium-carbon ratio of 4 for these steels to insure that normalizing, instead of annealing, may be used to develop satisfactory properties after fabrication.

Tensile Properties

The tensile properties of the steels mentioned in Fig. 1, normalized at 1700 degrees Fahr. (925 degrees Cent.), and with titanium-carbon ratios above 3, were excellent, except in a few instances with too high manganese contents, or where the bars were coarse-grained because the forging was finished too hot. Some examples of representative results, taken from numerous tests made in our laboratory, are reported in Table II, where they are arranged in the order of increasing titanium-carbon ratios for each group of steels.

Table II

Room-Temperature Tensile Test Results on Forged Bars of Steel Containing About 5 Per Cent Chromium and 0.5 Per Cent Molybdenum, Normalized at 1700 Degrees Fahr.

Steel No.	Chemical Analysis C	Mn	Si	Ti	Ti/C	Lbs. Per Sq. In. Yield Strength	Tensile Strength	Per Cent Elongation in 2 Inches	Reduction of Area
Normal manganese and silicon:									
1	0.07	0.40	0.34	0.20	2.9	111600	163600	14.0	52.5
2	0.08	0.30	0.22	0.30	3.8	34500	64300	38.0	85.4
3	0.09	0.58	0.30	0.49	5.4	29900	62200	38.0	79.5
4	0.07	0.41	0.32	0.54	7.7	31800	60300	41.5	84.7
High silicon:									
5	0.08	0.45	0.83	0.27	3.4	49000	84500	32.0	75.5
6	0.08	0.44	1.10	0.36	4.5	46000	75800	33.5	77.5
7	0.08	0.53	0.83	0.41	5.1	39900	70500	37.5	78.1
8	0.08	0.42	0.92	0.56	7.0	33900	58200	25.0	35.2
High manganese:									
9	0.12	1.42	0.66	0.58	4.8	98200	118700	16.5	61.5
10	0.09	0.89	0.33	0.53	5.9	33200	65000	39.5	80.0
11	0.08	0.98	0.30	0.60	7.5	30600	62300	39.0	80.2
12	0.13	0.76	0.52	1.28	9.8	37300	59600	26.7	48.2
High silicon and manganese:									
13	0.08	0.90	1.03	0.30	3.8	53300	93200	29.0	71.6
14	0.08	0.98	0.93	0.34	4.3	47100	79300	34.0	78.4
15	0.09	0.92	1.01	0.52	5.8	40600	67700	37.5	75.0
16	0.08	0.81	0.93	0.56	7.0	37200	63900	39.0	75.0

The excellent ductilities of most of the steels in Table II are noteworthy, and show that the properties as normalized may be satisfactory even with titanium-carbon ratios below 4. The specially high yield strengths of steels 1 and 9, with sufficient ductility to give reductions of area above 50 per cent, might make these compositions of interest for strong structural parts that cannot be heat treated conveniently by quenching. The properties of these steels, which are slightly air hardening, are modified appreciably by tempering as will be demonstrated later, but this would not be true of the other steels.

It is evident from Table II that silicon and manganese in these normalized 5 per cent chromium-molybdenum-titanium steels can be at least up to 1 per cent, even with titanium-carbon ratios slightly below 4, without appreciable impairment of the ductility, and with considerable advantage in yield strength for the higher contents.

Notched-Bar Impact Resistance

Impact tests on notched bars of the normalized steels listed in Table II, which are typical of numerous heats of the respective grades, gave much more divergent results than the tensile tests. A large number of Izod and Charpy tests, all with the standard V-notch, were made on these steels and similar compositions at room

temperature, but it seems unnecessary to report in detail more than the results on the typical steels of Table II. This is done in Table III, where the values given are chiefly averages of two to four results if they were in close agreement; or where the agreement was not close, the highest and lowest results are both reported.

Table III

Room-Temperature Izod Impact Test Results (Ft.-Lbs.) on Forged Bars of Steel Containing About 5 Per Cent Chromium and 0.5 Per Cent Molybdenum, Heat Treated as Noted

		Not Normalized			Normalized at 1700° F.			Quenched from 1950° F.	
Steel No.	Ti/C Ratio	Not Tempered	Tempered 1200° F.	Tempered 1350° F.	Not Tempered	Tempered 1200° F.	Tempered 1350° F.	Not Tempered	Tempered 1350° F.
Normal manganese and silicon:									
1	2.9	.....	.....	128	.....	151	.....	.	.....
2	3.8	171	173	180	.....	166	159	.	167
3	5.4	.....	.....	.....	114	.....	127	6	136
4	7.7	135	.....	132	13	.....	20–35	.	.....
High silicon:									
5	3.4	5	9	25	.....	185	185	.	25–50
6	4.5	50–180	.....	10–150	.....	.....	9–20	.	48–145
7	5.1	17–175	155	177	.....	12	15–95	.	57
8	7.0	6	.....	35	5	.....	15–24	9	7
High manganese:									
9	4.8	.....	.....	65–157	6	13	17	.	.....
10	5.9	.....	.....	.....	117	.....	125	7	137
11	7.5	122	152	169	.....	14–60	35	.	170
High silicon and manganese:									
13	3.8	.....	.....	40–107	.....	184	.....	.	.....
14	4.3	5	10–85	65–136	.....	187	35–184	.	25–75
15	5.8	34–110	.....	50–134	5–122	.....	118	4	120
16	7.0	25–50	20–60	20–40	.....	5	6	.	10

While the large spread in notched-bar impact values reported in Table III may appear to indicate an erratic response to this test in some of these titanium steels, most of the discrepancies can be explained by variations in the heat treatment or composition. Titanium-carbon ratios above about 6 should probably be avoided if notch-toughness is desired, especially in the normalized condition, or in the high silicon steels. With titanium-carbon ratios below 4, however, normalizing or tempering may be required for good notch toughness. Higher manganese helps the impact resistance of the high silicon steels, especially in the normalized condition. When these steels are quenched from 1950 degrees Fahr. (1065 degrees Cent.) they require tempering for good impact values, but even then quenching does not seem to be beneficial. Tempering at 1350 degrees Fahr. (730 degrees Cent.) was found to be generally preferable to 1200 or 1500 degrees Fahr. (650 or 815 degrees Cent.), though there was little difference usually between 1200 and 1350 degrees Fahr. (650 and 730 degrees Cent.). The time of tempering was 1.5 hours.

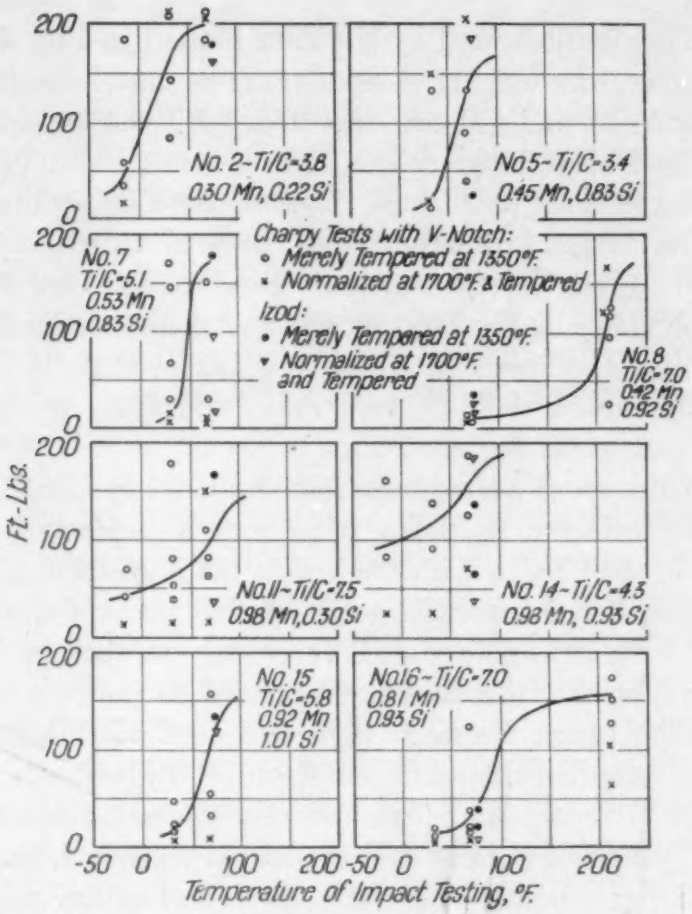


Fig. 2—Variation of Notched-Bar Impact Resistance with Temperature.

One important point concerning the impact resistance of these steels, that is not indicated in Table III, is that for good results the hot working of the steel must be performed in such a way as to produce a fine grain size. While the grain size may be coarsened by normalizing or other heat treatment at too high a temperature, it is practically impossible to refine it, in many of these steels, except by mechanical work. Avoiding a coarse grain size is essential for good notched-bar impact values, although the tensile results may not be affected. This is not difficult to accomplish with reasonably good forging or rolling practice, and by omitting any subsequent high temperature heat treatment.

The notch sensitivity of these steels is dependent, as in all other ferritic steels, on the temperature at which the impact is applied. The temperature of transition between high and low impact values seems to be not far below room temperature, or even above it, in some of the steels, especially those with higher silicon or titanium

contents. This is illustrated by the data plotted in Fig. 2, where all the low temperature impact results that we have obtained on the steels of Table II and III are reported. All the specimens represented in Fig. 2 were tempered at 1350 degrees Fahr. (730 degrees Cent.). Most of these tests were made by the Charpy method (with V-notch), and Fig. 2 illustrates our general experience that this method often gives erratic results. Good checks are much more often obtained here by the Izod method. All the pertinent Izod data from tempered specimens in Table III are plotted in Fig. 2 at a nominal temperature of 75 degrees Fahr. and agree fairly well with the Charpy data at 68 degrees Fahr.

The results from normalized and tempered specimens in Fig. 2 are generally inferior to those from merely tempered specimens, except only in steel No. 5 with the lowest titanium-carbon ratio. Normalizing was either detrimental, or at least of no benefit, in the other steels. The lines drawn in Fig. 2 refer only to the specimens tempered but not normalized, and are intended to indicate roughly the transition temperature between high and low notch-impact values. If it may be assumed that this temperature is about where the line crosses the 100 foot-pound line, the results can be summarized as in Table IV. Here it is clear that the lowest transition temperatures, and hence the least notch sensitivity, are found in the steels with titanium-carbon ratios of 5.1 or lower. It should be noted however that most of the data in Fig. 2 and Table IV apply to high silicon steels. Thus, although the conclusions drawn from Table III are supported by the low temperature data, there are not enough of these data on low silicon steels to warrant a definite conclusion that the titanium-carbon ratio should always be kept as low as 5.1 in them for satisfactory notch toughness. Certainly this does not appear to be true of steels with low silicon and over 0.75 per cent manganese.

Table IV

Summary of Low Temperature Impact Data on Specimens Merely Tempered, But Not Normalized, From Fig. 2

Steel No.	Chemical Composition		Temperature (°F.) at Which the Toughness-Temperature Curve Cuts the 100 Ft.-Lb. Line
	Si or Mn	Ti/C	
5	High Si	3.0	60
2	Normal	3.8	10
14	Both high	4.3	0
7	High Si	5.1	50
15	Both high	5.8	70
8	High Si	7.0	205
16	Both high	7.0	100
11	High Mn	7.5	70

Microstructures

In the course of our investigation of the notched-bar impact properties of these steels it was found that notch brittleness might occur not only when the grain size was too coarse from high temperature treatment, but also under some conditions in fine-grained material, especially with higher silicon. Thus the structural conditions for satisfactory toughness involve more than mere fine grain size, and an account of some experience along this line may be of interest, even though a complete explanation of all the structures has not been reached.

All these steels consist essentially of ferrite and titanium carbide (or carbo-nitride), except perhaps when quenched from 1700 degrees Fahr. (925 degrees Cent.) or higher, when a separate phase, possibly austenitic in nature, can generally be distinguished. The titanium carbide phase occurs in the form of fine angular crystals that can be shown in their proper size in a well-polished section, although after the repeated alternate polishing and etching that is required to bring out the ferrite structure free from "disturbed metal", these crystals become eroded and pitted so that they appear much larger than they should, giving the sections an abnormally dirty appearance. This should be kept in mind in connection with the photomicrographs shown here. Chromium steels require much more repolishing and re-etching than plain carbon steels to bring out the microstructure clearly, and etched sections of chromium steels with titanium therefore always look more or less dirty.

Figs. 3 and 4 show fine-grained structures of steel No. 10, with 0.89 per cent manganese and 5.9 times as much titanium as carbon, where the toughness was satisfactory in the annealed condition, although after high temperature heat treatment it was not, even though the grain size appears finer. Figs. 5 and 6 show likewise fine-grained structures in a higher silicon steel with a titanium-carbon ratio of 6, where heat treatment at 1700 degrees Fahr. (925 degrees Cent.) gave low impact values without grain coarsening.

Figs. 7 to 10, however, illustrate how the distorted structure due to cold working gives better toughness in these titanium steels, and how a coarse grain produced by high temperature normalizing gives lower toughness than the fine grain found after quenching the colder finished steel from the same high temperature, when both are tempered. Tempering at 1350 degrees Fahr. (730 degrees Cent.) or

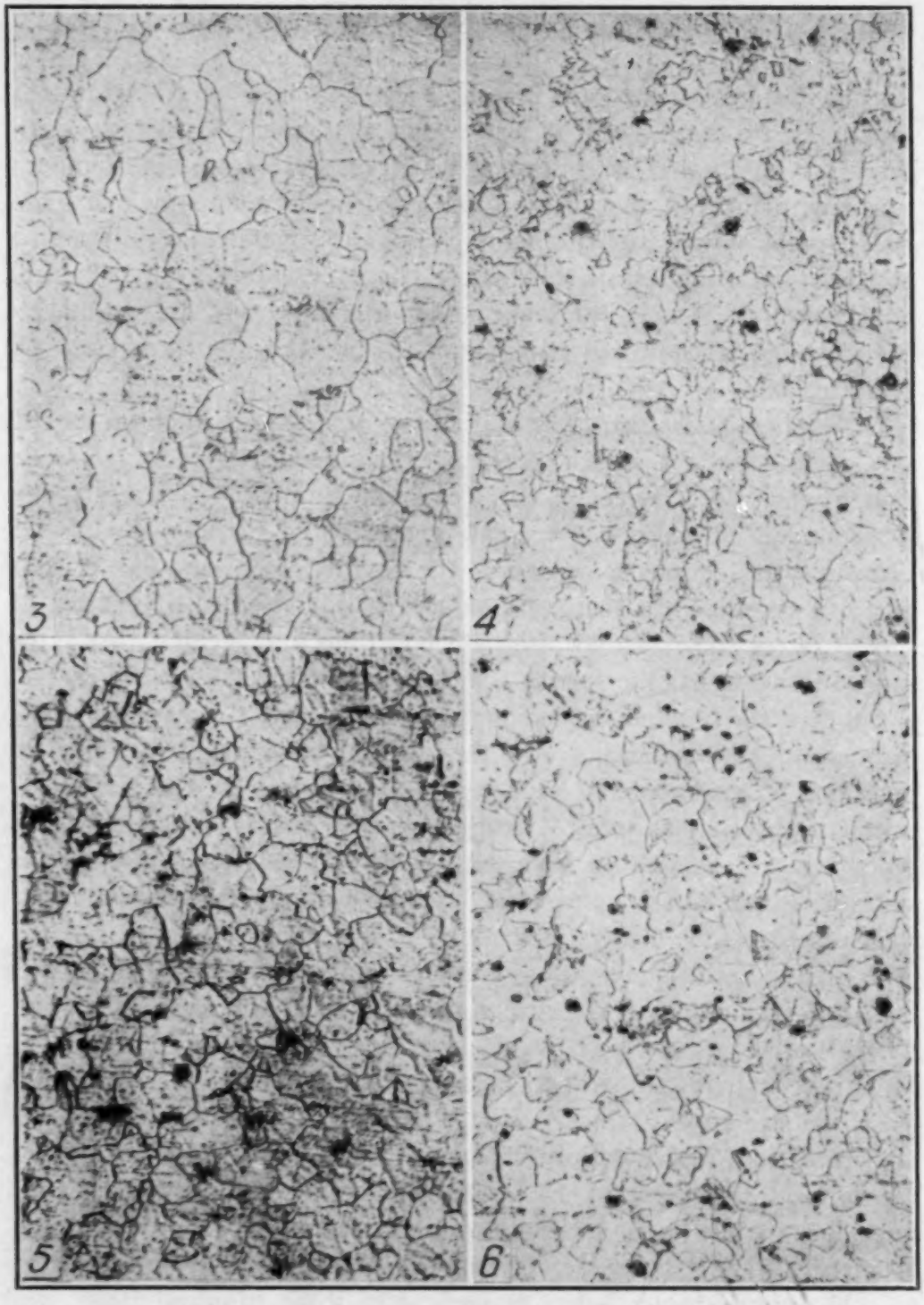


Fig. 3—Steel No. 10 with 0.89 Per Cent Manganese, 0.33 Per Cent Silicon, and Titanium/Carbon 5.9, Normalized at 1700 Degrees Fahr. and Annealed at 1550 Degrees Fahr. Izod impact value 120 foot-pounds. All specimens etched with nital. × 200.

Fig. 4—Structure of Same Steel as Fig. 3, Normalized at 1950 Degrees Fahr. with Only 20 Foot-Pounds Impact Value Though Finer Grained, Brinell Hardness 225.

Fig. 5—Structure of 5.00 Per Cent Chromium-Molybdenum-Titanium Steel with 0.67 Per Cent Manganese, 0.94 Per Cent Silicon, and Titanium/Carbon 6.0, Normalized at 1700 Degrees Fahr., Showing Ghosts in the Ferrite Grains. Izod impact value 22.

Fig. 6—Structure of Same Steel as Fig. 5 Quenched in Water from 1700 Degrees Fahr., Showing Two Phases. Izod impact value 10 foot-pounds.

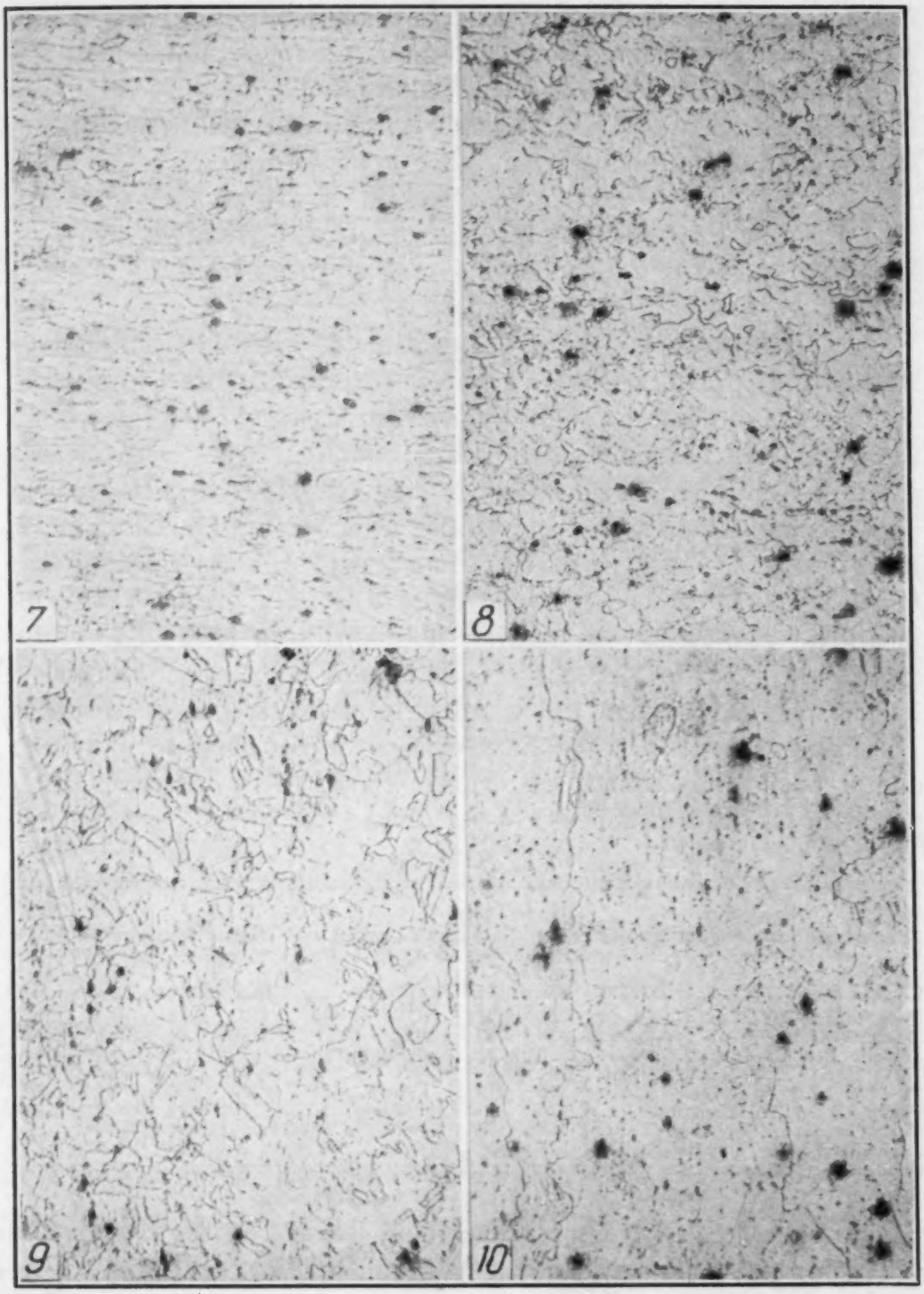


Fig. 7—Tough Structure of Steel No. 15, with 0.92 Per Cent Manganese, 1.01 Per Cent Silicon, and Titanium/Carbon 5.8, Not Heat Treated. Izod impact value 104 foot-pounds. All specimens etched with nital. × 200.

Fig. 8—Less Favorable Forged Structure of Same Steel as Fig. 7, with Only 47 foot-pounds impact value.

Fig. 9—Structure of Steel Like Fig. 7 After Quenching from 1950 Degrees Fahr. and Tempering at 1350 Degrees Fahr. with 120 Foot-Pounds Impact Value.

Fig. 10—Structure of Steel Like Fig. 8 After Normalizing from 1950 Degrees Fahr. and Tempering at 1350 Degrees Fahr. with 27 Foot-Pounds Impact Value.

less does not alter the microstructures of these steels. Their microstructures in general are not as informative in regard to the mechanical properties as in pearlitic steels, and a simple fracture inspection is often just as useful and certainly far simpler than a determination of the microstructure by metallographic means.

Effects of Phosphorus, Molybdenum, Nitrogen, and Aluminum Contents Higher than Normal

Phosphorus—Since it has been stated (11) that phosphorus may be used to improve the high temperature strength of steel in general, and especially of the 5 per cent chromium steels (12), its effects on the 5 per cent chromium-molybdenum-titanium steels were investigated by means of a few heats containing from 0.07 to 0.21 per cent phosphorus. The tensile properties of these steels were satisfactory, though with as much as 0.2 per cent phosphorus the strength was increased about 20 per cent and there was an appreciable drop in ductility. The air hardening after normalizing at 1700 to 1950 degrees Fahr. (925 to 1065 degrees Cent.) was unexpectedly increased by phosphorus, to about the same degree as with manganese up to 1 per cent. Values of over 200 Brinell were not obtained, however, except on cooling from 1950 degrees Fahr. (1065 degrees Cent.) and with phosphorus above 0.14 per cent.

Table V

Comparable Izod Impact Values in Foot-Pounds for Normal and Higher Phosphorus Steels

Class of Steel	Low Si and Mn		Higher Si		Higher Mn		Higher Si and Mn	
Phosphorus, Per Cent	0.02	0.073	0.02	0.073	0.02	0.08	0.02	0.071
Manganese, Per Cent	0.58	0.42	0.42	0.37	0.98	0.80	0.67	0.75
Silicon, Per Cent	0.30	0.27	0.92	1.01	0.30	0.34	0.94	1.04
Ti/C Ratio	5.4	5.5	7.1	6.7	7.3	7.7	6.0	6.1
Ft.-Lbs. Izod As-Forged	135	14	6	5	122	120	100	14
Tempered at 1350 Deg. Fahr.	170	159	35	9	169	135	71	32
Normalized at 1700 Deg. Fahr.	114	17	5	6	....	....	20–100	9
Normalized and Tempered	127	9–20	15–24	9	35	23	22	10

In regard to Izod impact resistance, Table V gives comparable values for steels containing about 0.02 and 0.075 per cent phosphorus respectively. Higher phosphorus contents seemed to lower the impact values too much to be useful in these titanium steels, especially with higher silicon.

It is concluded from comparisons such as appear in Table V that phosphorus contents up to 0.07 or 0.08 could be used only with

manganese around 0.80 per cent and with low silicon, on account of the tendency of phosphorus to promote notch sensitivity in this ferritic chromium steel with titanium. Possibly further work would be justified on this type of steel with higher than normal phosphorus, especially with titanium-carbon ratios of about 5 or less, and merely tempered after rolling or forging.

Molybdenum—Only two 5 per cent chromium-molybdenum-titanium steels were tested with more than about 0.6 per cent molybdenum, one having 0.34 per cent silicon and 0.92 per cent molybdenum, and the other 0.96 per cent silicon and 0.96 per cent molybdenum. Both had titanium-carbon ratios of about 7, and low manganese. Their tensile properties were about normal, and they did not harden on normalizing even from 1950 degrees Fahr. (1065 degrees Cent.), but the Izod impact values were low even when forged with special care and not normalized. No values above 26 foot-pounds were obtained with any heat treatment, and the fractures were mostly coarse. These steels were therefore not investigated further.

Nitrogen—The influence of up to 0.0287 per cent nitrogen was investigated in some of these 5 per cent chromium-molybdenum-titanium steels with about 1 per cent silicon, both normal and higher manganese respectively, and titanium-carbon ratios of 5 to 6. Neither the tensile properties nor the air hardening seemed to be affected by the higher nitrogen. In the steels with normal manganese contents the effect of nitrogen on the notched-bar impact resistance was negligible; and in the higher manganese steels, nitrogen increased from 0.016 per cent to 0.028 per cent had about the same effect on the impact values as reducing the titanium-carbon ratio from 5.1 to 4.4. That is, the steel then required tempering or normalizing for good impact resistance, instead of having high values as forged, and low after normalizing. This may be illustrated by the comparison

Table VI
Comparable Izod Impact Values of Steels with High Manganese and Silicon, and Different Nitrogen or Aluminum Contents

	Per Cent				
Class of Steel	0.012 N	0.016 N	0.028 N	0.1 Al Added	0.5 Al Added
Manganese, Per Cent	0.98	0.90	0.92	0.81	0.90
Silicon	0.93	1.04	1.11	0.93	1.02
Ti/C Ratio	4.4	5.1	5.0	7.5	7.0
Ft.-Lbs. Izod As-Forged	5	168	10	33	5
Tempered at 1350 Degrees Fahr.	112	106	145	33	7
Normalized at 1700 Degrees Fahr. and Tempered	184	13	181	6	5

in Table VI. From this it was concluded that nitrogen is not specially useful as a grain refiner in these 5 per cent chromium steels with titanium.

Aluminum—Since thorough deoxidation with aluminum is known to promote high impact values, especially at low temperatures (13), an extra large aluminum addition was tried in one of these steels with high silicon and titanium to see if the impact resistance would be improved. As shown in the right-hand columns of Table VI, however, the larger aluminum addition was detrimental rather than beneficial to the impact values. The tensile properties of both steels were satisfactory, and there was no air hardening in either.

Welding

The 5 per cent chromium-molybdenum-titanium steels are readily welded, and are suitable for welded construction if certain precautions are taken to insure that the final properties of the weld are satisfactory for the service conditions. When the 5 per cent chromium-molybdenum steel without titanium is welded the heat affected zone becomes hard and brittle unless annealed and cooled slowly (14). With an effective amount of titanium in the steel, this hardening may be more or less reduced, and even though it occurs it can be eliminated by normalizing with air cooling, instead of requiring slow cooling (10).

Table VII
Hardness and Ductility of 5 Per Cent Chromium-Molybdenum-Titanium Arc Weld Deposits, Not Heat Treated

Sample	Analysis of Weld Metal						Rockwell B Hardness		Duplicate Slow Bends —Not Notched—		Izod Value (Notched)
	C	Mn	Si	Cr	Mo*	Ti	Av.	Max.	Max. Degrees	Fracture	Av. Ft.-Lbs.
B	0.06	0.33	1.52	5.08	0.58	1.23	80	87	87–180	Coarse	2
C	0.04	0.37	1.31	4.34	0.43	0.72	81	82	154–180	Coarse	5
D	0.06	0.36	1.66	5.26	0.57	1.92	90	95	180–180	None	Cracked
E	0.12	0.42	1.17	4.74	0.41	0.45	102	105	158–180	Fine	2
F	0.06	0.33	0.43	4.36	0.43	0.30	100	105	148–180	Fine	10
G	0.07	0.33	0.82	5.28	0.58	0.64	88	99	180–180	None	6

*Note: Molybdenum was determined only on the welding rods used.

Weld deposits of 5 per cent chromium-molybdenum-titanium steel have been made by arc welding, using rods containing 2 to 4 per cent titanium, that were soft and ductile as welded, without any heat treatment, because sufficient titanium was retained in the deposits to stabilize them against air hardening. This is illustrated by

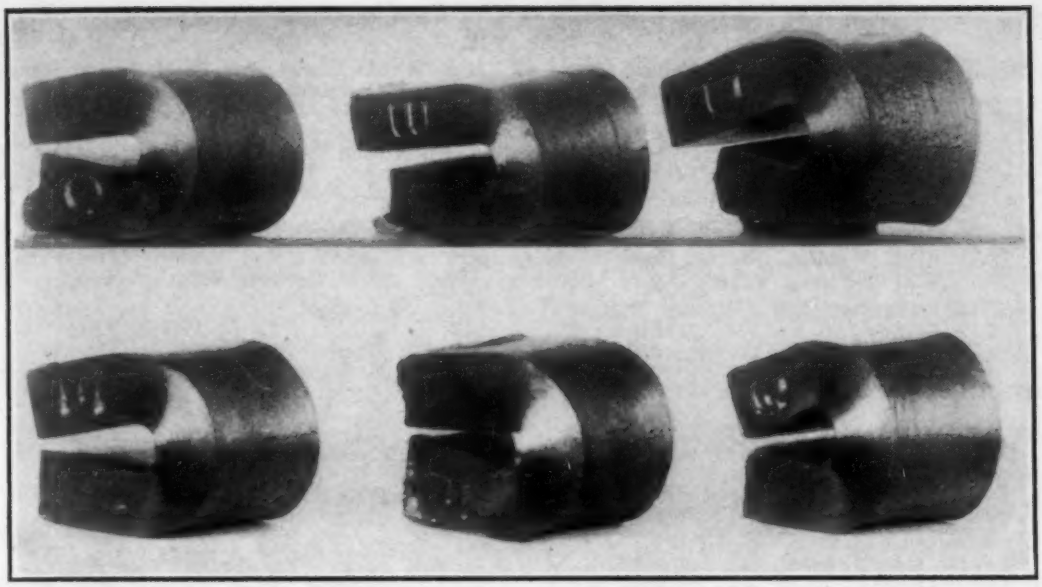


Fig. 11—Cold Bends of Small Specimens Machined from Arc-Weld Deposits of 5.00 Per Cent Chromium-Molybdenum-Titanium Steel, Not Heat Treated, Magnified About 2.5 Diameters to Show High Ductility and No Cracking.

the data in Table VII, and the photograph, Fig. 11, of bent weld-metal specimens, not heat treated.

The data in Table VII show that although these arc welds were soft and ductile without any heat treatment, their notch sensitivity was high. The Izod values were better in the weld metal with lower silicon and titanium, and could probably have been improved by heat treatment. These weld data, however, are chiefly of academic interest, since the high titanium rods with which they were made are not commercially available. Satisfactory results in welding this steel, using 18-8 columbium, or 25-20 rods, have been reported by Wyche (15), and such methods should no doubt be used at present. He considered the 5 per cent chromium-molybdenum-titanium steel to be essentially a "nonhardening base metal for welding", as indicated by the results of his tests.

Flash welding of the 5 per cent chromium-molybdenum-titanium is readily performed without any special difficulty. There always seems to be some hardening at the weld with this process, which is not corrected by tempering at 1350 degrees Fahr. (730 degrees Cent.), although normalizing at 1500 degrees Fahr. (815 degrees Cent.) or higher eliminates it (6), (16). Hardness data on sections of flash welds of a few of these steels are presented graphically in Fig. 12. These results were obtained after tempering at 1350 degrees Fahr. (730 degrees Cent.), which evidently was high

Table VIII

Results of Charpy V-Notch Impact Tests of Flash Welds Between ¾-Inch Round Bars of 5 Per Cent Chromium-Molybdenum-Titanium Steels

		Welded as Forged Tempered at 1350 Degrees Fahr. —After Welding—			Normalized at 1700° Fahr. Before Welding— Not Tempered After —Welding—			Normalized at 1700° Fahr. Before Welding— Tempered at 1350 Degrees Fahr. After Welding—		
Steel No.	Ti/C Ratio	Time of Welding Seconds	Impact Value Ft.-Lbs.	Fracture	Time of Welding Seconds	Impact Value Ft.-Lbs.	Fracture	Time of Welding Seconds	Impact Value Ft.-Lbs.	Fracture
Normal manganese and silicon:										
2	3.8	6	7.2	Fine	.	...	....	7	19	Fine
3	5.4	.	...	....	5	2.9	Fine	5	40	Fine
4	7.7	.	...	....	.	1.4	Medium	.	48	Medium
High silicon:										
5	3.0	6	2.9	Fine	.	...	....	8	8	Fine
7	5.1	5	12.3	Fine	.	...	....	4	12	Fine
8	7.0	.	...	....	6	2.9	Coarse	7	4	Coarse
High manganese:										
10	5.9	.	...	....	5	2.2	Fine	6	144	Fine
11	7.5	6	14.5	Fine	.	...	....	5	23	Fine
High silicon and manganese:										
14	4.3	8	3.6	Fine	.	...	....	7	29	Fine
15	5.8	.	...	....	4	2.2	Medium	6	15	Medium
16	7.0	6	3.6	Coarse	.	...	....	8	4	Coarse

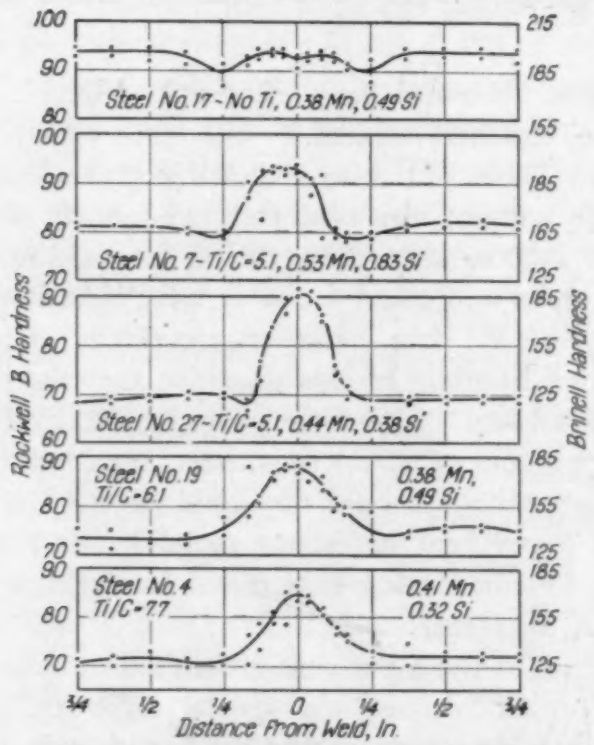

Steel No. 17 - No Ti, 0.38 Mn, 0.49 Si
Steel No. 7 - Ti/C = 5.1, 0.53 Mn, 0.83 Si
Steel No. 27 - Ti/C = 5.1, 0.44 Mn, 0.38 Si
Steel No. 19 Ti/C = 6.1
0.38 Mn, 0.49 Si
Steel No. 4 Ti/C = 7.7
0.41 Mn 0.32 Si
Rockwell B Hardness
Brinell Hardness
Distance From Weld, In.
¾ ½ ¼ 0 ¼ ½ ¾

Fig. 12—Hardness of Flash Welds of 5.00 Per Cent Chromium-Molybdenum Steels. Tempered at 1350 degrees Fahr.

enough to modify appreciably the weld-hardness of the titanium-free steel. All of the steels are seen to reach about the same maximum hardness at the weld, though the titanium steels are softer where not welded.

Charpy impact tests of flash welds were made at room temperature to reveal their toughness, using standard V-notched specimens with the notch cut cross-wise at the weld line. The results are reported in Table VIII, where the various conditions of heat treatment are noted. All these welds were made with the same equipment by the same operator and supposedly under uniform conditions.

The data in Table VIII indicate marked notch sensitivity in most of these flash welds, with a few exceptions when the specimens were normalized before welding and tempered ("stress-relieved") afterward. The high silicon, high titanium steels were coarse-grained and brittle in that condition. The toughness of the tempered specimen of steel No. 10 was checked with another steel containing 1.06 per cent manganese and with a titanium-carbon ratio of 6.6, which had an impact value of 131 foot-pounds when flash-welded and tested under the same conditions. There seems to be some indication that the flash welds made more quickly were tougher. A notable improvement in toughness was certainly produced by tempering the welds of the normalized specimens, and it is possible therefore that the toughness of the arc welds of Table VII might also have been improved by a similar treatment. The time of tempering was 1.5 hours, the same as for the impact specimens not welded.

Oxidation at High Temperature

Several series of oxidation tests were made on a few 5 per cent chromium-molybdenum-titanium steels at 1100 to 1700 degrees Fahr. (595 to 925 degrees Cent.) by heating small spool-shaped specimens in an electric muffle furnace, with small openings at top and bottom to permit some circulation of air. The tests were all made in triplicate, and for a period of 1000 hours, all the specimens being heated

Table IX
Average Scale Losses, in Grams Per Square Inch, During 1000 Hours at the Temperatures Noted

Steel	Chemical Composition									Temperature of Test — Degrees Fahr.			
	C	Mn	Si	Cr	Mo	Ti	Al	Cb	Ti/C	1100	1300	1500	1700
17	0.12	0.38	0.49	5.42	0.47	....	....	....	....	0.10	0.13	0.46*	7.07
18	0.09	0.40	0.36	4.99	0.56	....	....	0.59	....	0.11	0.60	2.20*	7.58
19	0.08	0.38	0.40	5.16	0.50	0.49	....	....	6.1	0.12	0.11	0.41	6.40
20	0.08	1.42	0.66	4.86	0.10	0.75	....	....	9.4	0.08	0.19	0.28	6.57
21	0.12	0.28	1.42	5.20	0.50	....	....	....	....	0.03	0.01	0.03	1.59*
22	0.19	0.26	1.26	5.03	0.51	....	0.84	....	....	0.01	0.01	0.02	0.92*
23	0.16	0.79	1.22	5.70	0.44	0.90	....	....	5.6	0.03	0.01	0.09	(0.31?)
24	0.15	0.74	1.49	5.50	0.55	1.65	....	....	11.0	0.01	0.00	0.02	0.08

*Signifies that the surface was badly pitted after testing.

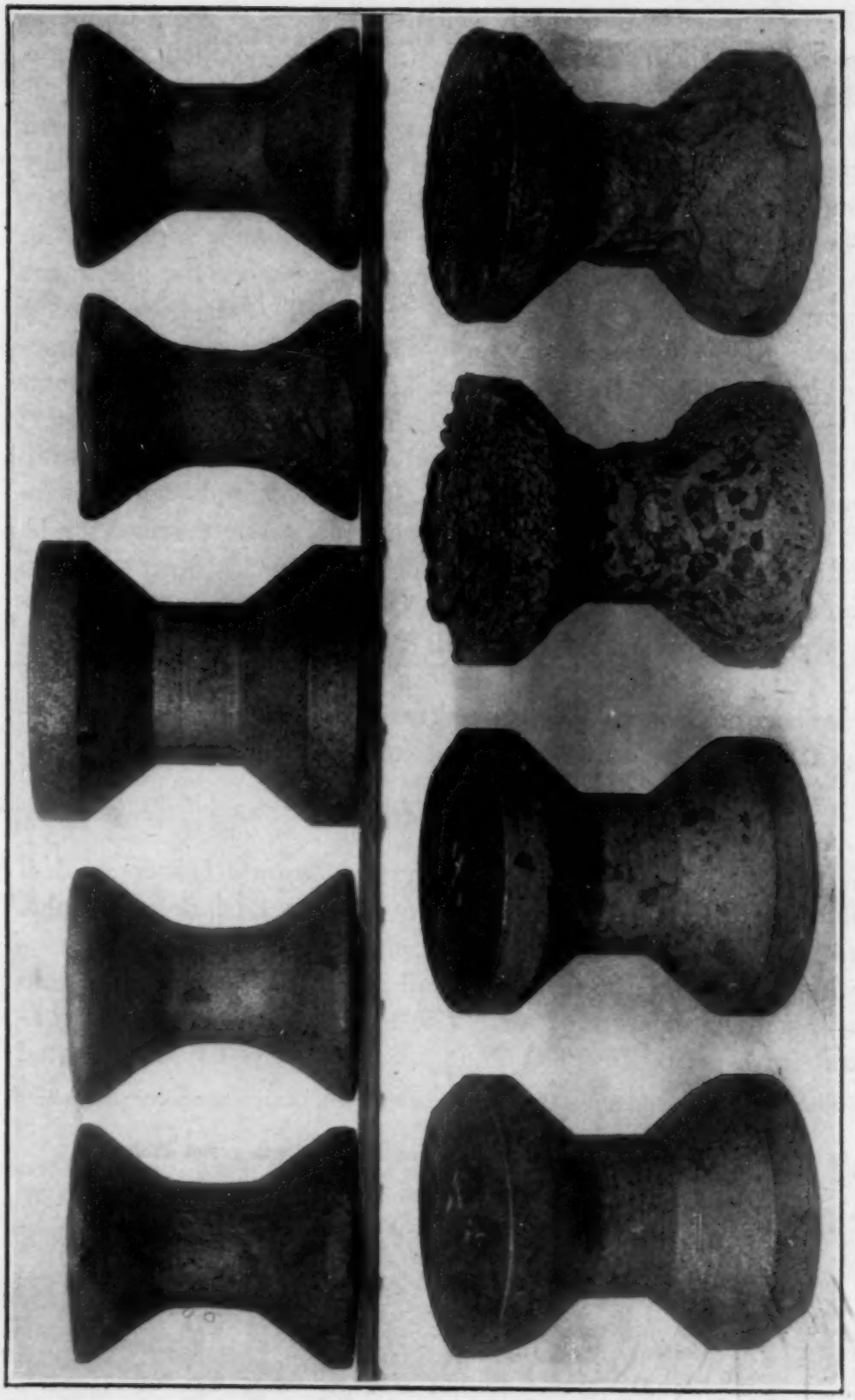


Fig. 13—Oxidation Test Specimens After 1000 Hours at 1700 Degrees Fahr., Cleaned by Sand-Blasting, and Magnified About 2.3 Diameters. Top row, left to right: Steel Nos. 20 (manganese-titanium) and 17 (no titanium), then an unoxidized specimen to show the original size, then Nos. 18 (columbium) and 19 (titanium). Lower row, left to right: Steel Nos. 24 (silicon-titanium), 23 (silicon-titanium), 21 (silicon), and 22 (silicon-aluminum).

together at each temperature. The steels used for these tests were not from the same heats listed in Table II, and included some steels without titanium for comparison. Their compositions are reported together with the results of the tests in Table IX. Each result is the average of three values for loss in weight per square inch of original surface, obtained from the difference between the original weight and the weight after oxidation and removal of the scale by sand blasting.

The data in Table IX show that 1 per cent higher silicon in these steels makes them much more resistant to oxidation at high temperatures. In both the low silicon and high silicon steels, titanium also had a beneficial effect in reducing the oxidation loss, though it was not as effective as silicon. The appearance of the specimens after the 1700-degree Fahr. (925-degree Cent.) treatment is illustrated by Fig. 13, where the pitted surfaces of samples 21 and 22 are noticeable. Similar pitting also occurred in the other two titanium-free steels 17 and 18 in testing at 1500 degrees Fahr. (815 degrees Cent.), as noted in Table IX, but was not found in any of the titanium steels. The titanium steels, therefore, not only oxidize less than steels of similar silicon content without titanium, but also oxidize more uniformly, without the formation of pits that might cause failure in a structure with only a moderate total oxidation loss. High manganese, as in steel No. 20, does not seem to affect the rate of oxidation.

Similar results were reported by Habart (17) on some of these steels, but greater losses were found at 1700 degrees Fahr. (925 degrees Cent.) in those tests than in the tests here reported. This was probably due to the fact that the present tests were made in the winter, when the air was dry, and the tests reported by Habart were made in the summer when more moisture was present in the air. Oxidation losses are known to be seriously affected by the amount of water vapor in the atmosphere (18), and this point should always be considered in comparing results obtained at different times.

Time for Rupture Under Load at High Temperature

Tests of this kind, commonly called "stress-rupture tests", were made on an extensive series of specimens from four commercial 5 per cent chromium-molybdenum-titanium steels, both in the annealed and normalized conditions, and at 1000 and 1200 degrees Fahr. (540 and 650 degrees Cent.). The results of the tests on these heats were

in sufficiently close agreement to establish fairly reliable stress-rupture time curves for this class of steel, with the usual commercial titanium-carbon ratios, at the temperatures noted. Tests to show the effects of variations in composition or heat treatment have been made so far only at 1200 degrees Fahr. (650 degrees Cent.), and although complete curves have not yet been established for these

Table X

Rupture Strengths at 1000 and 1200 Degrees Fahr. of Commercial 5 Per Cent Chromium-Molybdenum-Titanium Steels

Steel No.	4	25	26	27
Carbon, Per Cent	0.07	0.06	0.06	0.10
Manganese	0.41	0.40	0.43	0.44
Silicon	0.32	0.48	0.25	0.38
Chromium	5.46	5.43	5.72	4.98
Molybdenum	0.57	0.49	0.48	0.50
Titanium	0.54	0.48	0.41	0.51
Titanium-Carbon Ratio	7.7	8.0	6.8	5.1
Specimens Annealed at 1550 Degrees Fahr.:				
Brinell Hardness	112	124	124	126
A.S.T.M. Grain Size	1, 6–8	7–8	6–8	8
Stress for Rupture in 1000 Hours at 1000 Degrees Fahr., psi.	21000	20000	21000	22000
Stress for Rupture in 10,000 Hours at 1000 Degrees Fahr., psi.	12250	12750	13500	14500
Stress for Rupture in 1000 Hours at 1200 Degrees Fahr., psi.	5400	5400	5800	5000
Stress for Rupture in 10,000 Hours at 1200 Degrees Fahr., psi.	3950	3750	4200	3300
Specimens Normalized at 1750 Degrees Fahr.:				
Brinell Hardness	112	117	120–140	187
A.S.T.M. Grain Size	6–8, 1–8	5–7	4–7	7–8
Stress for Rupture in 1000 Hours at 1000 Degrees Fahr., psi.	23500	22000	23500	30000
Stress for Rupture in 10,000 Hours at 1000 Degrees Fahr., psi.	11250	11500	13000	22000
Stress for Rupture in 1000 Hours at 1200 Degrees Fahr., psi.	6300	5400	5400	4900
Stress for Rupture in 10,000 Hours at 1200 Degrees Fahr., psi.	4600	3750	3600	3000

variations, sufficient data can be reported to indicate the possibility of effecting considerable improvement in the resistance of this kind of steel to failure at high temperatures by controlling the composition and previous treatment.

The analyses and test results on the four commercial heats are reported in Table X, and the rupture-time data are plotted in Figs. 14 and 15.

All the specimens used for the tests reported in Table X and Figs. 14 and 15 showed very ductile fractures, the only reduction of area value below 50 per cent having been found in the specimen of steel No. 27 that broke after 1980 hours at 1200 degrees Fahr. (650 degrees Cent.) where the elongation was 37 per cent and the reduction of area 44.9 per cent. There was some intercrystalline crack-

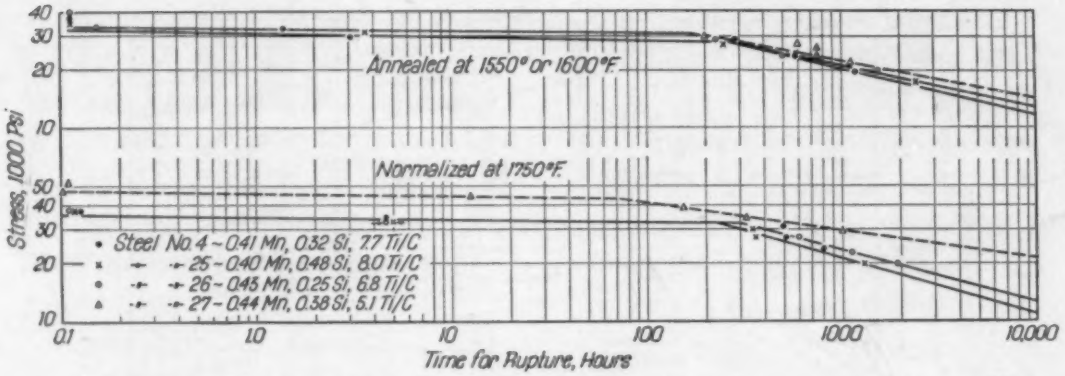

Annealed at 1550° or 1600°F.
Normalized at 1750°F.
Steel No. 4 - .041 Mn, 0.32 Si, 7.7 Ti/C
25 - 0.40 Mn, 0.48 Si, 8.0 Ti/C
26 - 0.45 Mn, 0.25 Si, 6.8 Ti/C
27 - 0.44 Mn, 0.38 Si, 5.1 Ti/C
Stress, 1000 Psi
Time for Rupture, Hours

Fig. 14—Stress Versus Time-for-Rupture Curves at 1000 Degrees Fahr. for Steels 4, 25, 26 and 27.

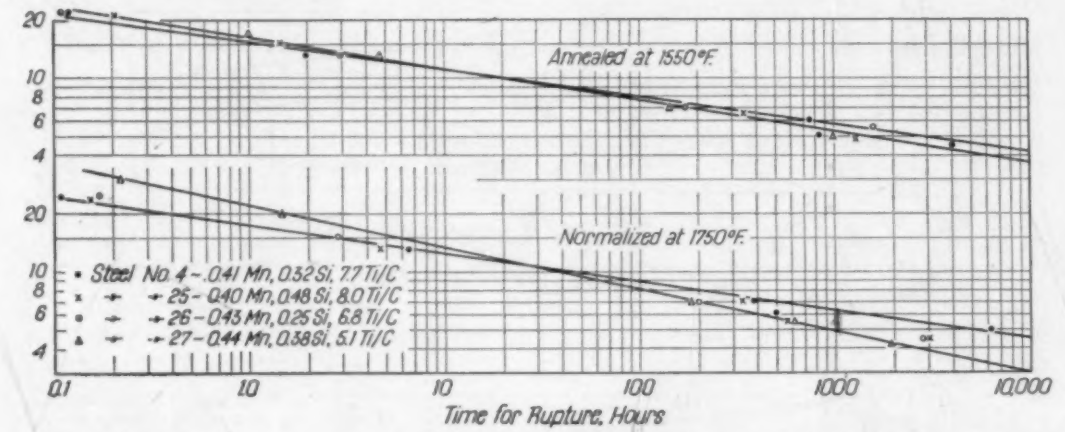

Annealed at 1550°F.
Normalized at 1750°F.
Steel No. 4 - .041 Mn, 0.32 Si, 7.7 Ti/C
25 - 0.40 Mn, 0.48 Si, 8.0 Ti/C
26 - 0.45 Mn, 0.25 Si, 6.8 Ti/C
27 - 0.44 Mn, 0.38 Si, 5.1 Ti/C
Time for Rupture, Hours

Fig. 15—Stress Versus Time-for-Rupture Curves at 1200 Degrees Fahr. for Steels 4, 25, 26 and 27.

ing in this steel, but it was absent or negligible in the others. Coarse structures in general gave longer rupture-time periods than the finer structures. The greater hardness of steel No. 27, after normalizing, possibly due to its lower titanium-carbon ratio, is reflected in the higher strength values at 1000 degrees Fahr. (540 degrees Cent.), but this effect did not persist at 1200 degrees Fahr. (650 degrees Cent.).

A few stress-rupture tests have been made at 1200 degrees Fahr. (650 degrees Cent.) on some of the 5 per cent chromium-molybdenum-titanium steels with compositions modified like those that have already been described, and heat treated by tempering 1.5 to 2 hours at 1350 degrees Fahr. (730 degrees Cent.) as well as by normalizing or annealing at higher temperatures. Most of these steels were forged from small ingots, and care was taken not to forge them too hot. The results of these tests are plotted in Figs. 16 and 17. The full lines drawn on each one of these charts are the same as

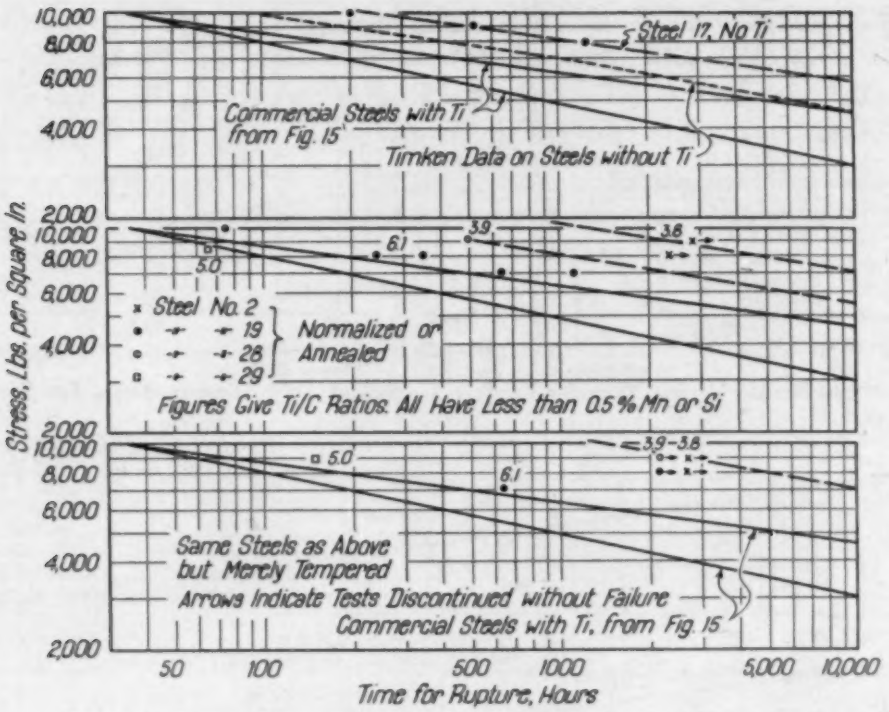

Steel 17, No Ti
Commercial Steels with Ti from Fig. 15
Timken Data on Steels without Ti
Steel No. 2
19
28
29
Normalized or Annealed
Figures Give Ti/C Ratios. All Have Less than 0.5 % Mn or Si
Same Steels as Above but Merely Tempered
Arrows Indicate Tests Discontinued without Failure
Commercial Steels with Ti, from Fig. 15
Stress, Lbs. per Square In.
Time for Rupture, Hours

Fig. 16—Stress Versus Time-for-Rupture Curves at 1200 Degrees Fahr. for Steels 2, 17, 19, 28 and 29.

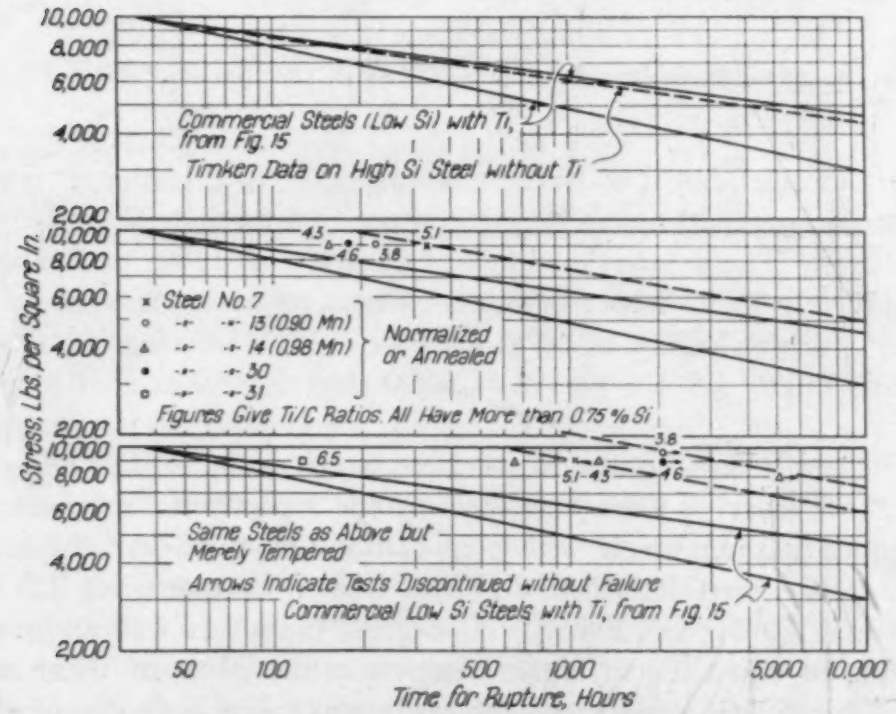

Commercial Steels (Low Si) with Ti, from Fig. 15
Timken Data on High Si Steel without Ti
Steel No. 7
13 (0.90 Mn)
14 (0.98 Mn)
30
31
Normalized or Annealed
Figures Give Ti/C Ratios. All Have More than 0.75 % Si
Same Steels as Above but Merely Tempered
Arrows Indicate Tests Discontinued without Failure
Commercial Low Si Steels with Ti, from Fig. 15
Stress, Lbs. per Square In.
Time for Rupture, Hours

Fig. 17—Stress Versus Time-for-Rupture Curves at 1200 Degrees Fahr. for High Silicon Steels 7, 13, 14, 30 and 31.

those shown in Fig. 15 to represent the spread of average results at 1200 degrees Fahr. (650 degrees Cent.) for the normalized commercial steels. The relation of the new data in Figs. 16 and 17 to the average values for commercial steel can be readily appreciated in this way. At the top of each of these figures a comparison with annealed 5 per cent chromium-molybdenum steel without titanium is presented, using data published by Timken (6) as well as a few results on steel No. 17 of Table IX. The dotted lines representing the Timken data are of course drawn as they were published, but the other broken lines are drawn parallel to the average slope for the commercial titanium steels. It is admitted that not enough data are presented to establish these broken lines accurately, but by means of them a convenient, though rough, comparison with the data on the commercial steels of Table X can be arrived at. The comparable

Table XI

Estimated Rupture Strengths at 1200 Degrees Fahr. of Experimental 5 Per Cent Chromium-Molybdenum-Titanium, and Other 5 Per Cent Chromium-Molybdenum Steels

Low Silicon Steels (Fig. 16):	Average Com.	Timken No Ti	17	2	28	29	19
Manganese, Per Cent	0.40–0.44	0.5 max.	0.38	0.30	0.37	0.49	0.38
Silicon	0.25–0.48	0.5 max.	0.49	0.22	0.31	0.48	0.40
Ti/C Ratio	5.1 –8.00	None	None	3.8	3.9	5.0	6.1
Normalized or Annealed:							
Stress for Rupture in 1000 Hours, psi.	5450	7000	8100	10000	8000	5000	6500
Stress for Rupture in 10,000 Hours, psi.	3770	5000	5800	7000	5500	3500	4500
Tempered at 1350 Degrees Fahr.:							
Stress for Rupture in 1000 Hours, psi.	....	....	....	10100	10000	6500	7500
Stress for Rupture in 10,000 Hours, psi.	....	....	....	7000	7000	4500	....
High Silicon Steels. (Fig. 17):		**Timken No Ti**	**13**	**14**	**30**	**7**	**31**
Manganese, Per Cent		0.5 max.	0.90	0.98	0.42	0.53	0.32
Silicon		1–2	1.03	0.93	0.77	0.83	0.90
Ti/C Ratio		None	3.8	4.3	4.6	5.1	6.5
Normalized or Annealed:							
Stress for Rupture in 1000 Hours, psi.		6000	7000	6000	6500	7400	....
Stress for Rupture in 10,000 Hours, psi.		4400	....	....	....	....	....
Tempered at 1350 Degrees Fahr.:							
Stress for Rupture in 1000 Hours, psi.		....	10100	9000	10000	9000	6500
Stress for Rupture in 10,000 Hours, psi.		....	7400	6000	7000	6000	....

data obtained in this way are presented in Table XI. The arrows in Figs. 16 and 17 indicate that the specimens were not broken when the tests were stopped. The figures near the plotted points indicate the titanium-carbon ratio. Most of the specimens that ruptured showed excellent ductility, the lowest values being 18 per cent elongation and 71 per cent reduction of area for the specimen that broke after the longest period.

The results in Figs. 16 and 17 and Table XI show that considerable variation is possible in the stress-rupture values of the 5 per cent chromium-molybdenum-titanium steels at 1200 degrees Fahr. (650 degrees Cent.), depending on the titanium-carbon ratio and the heat treatment applied before loading. In the steels of normal silicon content, these values with titanium-carbon ratios of 5 to 8 are lower than those of the 5 per cent chromium-molybdenum steels without titanium, although they may be improved to almost equal the latter if merely tempered instead of being used in the annealed or normalized condition. It should be remembered that the titanium steel does not need to be annealed after hot working as the titanium-free steel does, and merely tempering at 1350 degrees Fahr. (730 degrees Cent.) will generally reduce its hardness to 200 Brinell or lower. With titanium-carbon ratios of 3.8 or 3.9 however, the stress-rupture values at 1200 degrees Fahr. (650 degrees Cent.) are definitely higher than in the steel with no titanium, and of course much higher than with titanium-carbon ratios above 5. In these steels, with low titanium-carbon ratios, tempering seems to give about the same stress-rupture values as normalizing or annealing, though there may be a slight improvement with tempering.

With over 0.75 per cent silicon, the normalized or annealed titanium steels have no lower stress-rupture values than the steel without titanium, which is inferior to the low silicon steel in this respect according to the Timken data (6). Tempering the high silicon-titanium steels at 1350 degrees Fahr., instead of annealing, however, gives a marked improvement in the stress-rupture values at 1200 degrees Fahr. (650 degrees Cent.), even with titanium-carbon ratios of 3.8 to 5. A ratio above 5.1 gave an inferior result. Manganese between 0.90 and 1 per cent is compatible with good stress-rupture values at 1200 degrees Fahr. (650 degrees Cent.) in the 5 per cent

Table XII

Izod Impact Resistance of Stress-Rupture Specimens Under Load at 1200 Degrees Fahr. at Least 2000 Hours

Steel No.	Per Cent Mn	Per Cent Si	Per Cent Ti/C	Heat Treatment Before Stress-Rupture Test Degrees Fahr.	Ft.-Lbs. Izod Impact Resistance at Room Temperature
2	0.30	0.22	3.8	Annealed 1550	169–166–169
2	0.30	0.22	3.8	Tempered 1350	176–171–175
28	0.37	0.31	3.9	Tempered 1350	55–125
19	0.38	0.40	6.1	Tempered 1350	169–163–173
13	0.90	1.03	3.8	Tempered 1350	90–132
14	0.98	0.93	4.3	Tempered 1350	40–160–105
30	0.42	0.77	4.6	Tempered 1350	150–160

chromium-molybdenum-silicon-titanium steel, though it does not seem to be specially advantageous in this respect.

The rupture test specimens that were not tested to failure were machined to square Izod notched bars and tested by impact at room temperature. Some of the corners of these impact specimens were slightly rounded because of the size of the stress-rupture bars from which they were made, but this apparently had a negligible influence on the impact results. These results are given in Table XII.

The results in Table XII show that these steels do not become brittle after long service under load at 1200 degrees Fahr. (650 degrees Cent.) and their toughness and excellent stress-rupture values are evidence of satisfactory structural stability at that temperature.

Tensile Properties, Hardness, etc., When Merely Tempered

Since the stress-rupture results as well as the impact test results indicated that the properties of the 5 per cent chromium-molybdenum-titanium steels, especially with high titanium or silicon, may be better when they are merely tempered at 1350 degrees Fahr. (730 degrees Cent.) after hot working instead of being annealed or normalized at a higher temperature, it is important to determine the tensile and other properties in the merely tempered condition. This has been done on a restricted series of steels, including several found to have good stress-rupture values, the elastic properties being determined from carefully plotted stress-strain diagrams. The hardness and impact values in the same condition are also reported, along with the tensile properties, in Table XIII.

Table XIII

Tensile and Other Properties of 5 Per Cent Chromium-Molybdenum-Titanium Steels Merely Tempered After Hot Working

Steel No.	1	28	29	9	13	30	31
Manganese, Per Cent	0.40	0.37	0.49	1.42	0.90	0.42	0.32
Silicon	0.34	0.31	0.48	0.66	1.03	0.77	0.90
Ti/C Ratio	2.9	3.9	5.0	4.8	3.8	4.6	6.5
Yield Strength, 0.01% Offset	56500	81500	43000	76400	93100	52000	40900
Yield Strength, 0.2% Offset	69800	100900	47900	88600	106400	59600	46000
Tensile Strength	87700	111100	71400	102900	116200	80200	69900
Elongation in 2 Inches, Per Cent	22.5	18.5	29.0	24.0	21.0	31.5	30.0
Reduction of Area, Per Cent	71.6	70.0	77.5	64.5	70.2	79.4	77.9
Brinell Hardness	192	229	133	217	232	170	133
Rockwell B Hardness	94	100	83	96	102	90	81
Izod Impact, Ft.-Lbs.	128	121	173	11	40–107	134	40–155

According to the properties reported in Table XIII, tempering at 1350 degrees Fahr. (730 degrees Cent.) gives satisfactory duc-

tility and impact resistance in all these steels, except for the impact value of the high manganese steel No. 9. The strength is lower in this steel, and in No. 1, in the tempered condition than in the same steels normalized at 1700 degrees Fahr. (925 degrees Cent.) as reported in Table II. The hardness exceeds 200 Brinell slightly in some of the steels with titanium-carbon ratios below 4, but the ductility and impact values should be high enough in all except possibly No. 9 to enable them to withstand successfully all normal service conditions.

General Discussion of Results

Although no creep data are yet available on the steels of modified composition or heat treatment, the information here presented throws considerable light on possible reasons for the discrepancies in the published data in Table I. The strength of the 5 per cent chromium-molybdenum-titanium steel at high temperature, as well as its notch sensitivity in certain conditions of heat treatment, may be greatly affected by the titanium-carbon ratio. According to present specifications, any ratio above 4 is permissible, yet the high temperature properties of steels in which this ratio is around 4 to 5 may be quite different from those of steels with enough titanium to give a ratio of 7 or 8. This is believed to be the most likely reason for the discrepancies noted in Table I.

The properties reported for this steel have suffered not only from using too high a titanium-carbon ratio in it, but also from the custom, derived from experience with the air hardening 5 per cent chromium-molybdenum steel without titanium, of annealing the test specimens above 1500 degrees Fahr. (815 degrees Cent.). According to the data here presented, this seems to be unnecessary, and sometimes undesirable for the titanium steels in comparison with a tempering treatment at about 1350 degrees Fahr. (730 degrees Cent.).

The usual specifications for manganese and silicon in this steel limit these elements to 0.5 per cent although as has been shown by Habart (17) and others (5), (6), (7) the resistance to oxidation at high temperature is greatly improved by silicon contents around 1 per cent. These higher silicon steels have not been widely adopted however, probably because with the unrestricted upper limit for the titanium-carbon ratio and the generally used annealing heat treatment, the impact values and high temperature strength may be com-

paratively poor. It is here shown that with manganese contents around 1 per cent, titanium-carbon ratios below 5, and heat treatment restricted to tempering at 1350 degrees Fahr., much more satisfactory properties can be obtained in the 5 per cent chromium-molybdenum-silicon-titanium steels, so that the favorable oxidation resistance conferred by silicon could be utilized together with the freedom from air hardening due to titanium.

The specified low limit of 4 for the titanium-carbon ratio in these steels is probably unnecessary, since air hardening is negligible with slightly lower ratios, and titanium is specified only for preventing such hardening. For all practical purposes it is not necessary that air hardening be completely eliminated, but merely that it be restricted to a harmless degree. With too high titanium-carbon ratios, however, definitely inferior high temperature strength and greater notch sensitivity are experienced, so that a maximum limit should be specified for this ratio and not merely for the titanium content as is done at present. To avoid unnecessary difficulty in production due to establishing too close limits for titanium, the lower limit could be reduced, and specifying titanium-carbon ratios between 3.5 and 5.5, or even 3 to 6, would be preferable to the present practice tolerating anything between 4 and 10 or more.

Conclusions

The chief conclusions to be drawn from this investigation are the two following:

1. With titanium-carbon ratios between 3.5 and 5.5 in the 5 per cent chromium-molybdenum-titanium steel better notch toughness and high temperature rupture strength are obtained than with higher titanium, together with satisfactory ductility and restriction of air hardening.

2. Silicon around 1 per cent in these steels of 3.5 to 5.5 titanium-carbon ratio gives superior resistance to oxidation, and need not impair the notch toughness, high temperature rupture strength, or other properties if the manganese content is about 0.75 to 1 per cent and the steel is merely tempered at about 1350 degrees Fahr. (730 degrees Cent.) without annealing or normalizing, after hot working at a reasonably low temperature so as to avoid a coarse grain size.

Other minor conclusions can be drawn as follows:

3. Although titanium-carbon ratios above 6 may be required to eliminate all air hardening at 1950 degrees Fahr. (1065 degrees

Cent.), air hardening at 1700 degrees Fahr. (925 degrees Cent.) is suppressed by titanium-carbon ratios above 4. With ratios above 3, air hardening is not over 250 Brinell on cooling from 1700 degrees Fahr. (925 degrees Cent.) (except with high manganese and low silicon), or not over 160 Brinell with above 0.75 per cent silicon.

4. In these steels with silicon above 0.75 per cent, the transition temperature from low to high notch sensitivity is between room temperature and zero with titanium-carbon ratios below 5.5, and above room temperature with higher titanium-carbon ratios.

5. Higher than normal phosphorus, molybdenum, and aluminum in these titanium steels promote notch sensitivity, but about 0.08 per cent phosphorus might be tolerated with manganese up to 0.80 per cent and low silicon, or with a low titanium-carbon ratio. Nitrogen tends to reduce the effective titanium-carbon ratio.

6. The 5 per cent chromium-molybdenum-titanium steel may be welded without hardening to above 200 Brinell if the welds are tempered at 1350 degrees Fahr. (730 degrees Cent.). The notch sensitivity of welds tends to be high, but flash welds with fairly good impact values can be made if the silicon content is below 0.60 per cent or the manganese above 0.70 per cent, the steel is normalized before welding and tempered at 1350 degrees Fahr. (730 degrees Cent.) afterward, and the welding current is not applied too long, or not over 6 seconds for ¾-inch bars.

7. Oxidation resistance is improved slightly by titanium, not by high manganese, and greatly by more than 1.2 per cent silicon. The pitting type of scaling that occurs with moderate rates of oxidation was not found on titanium steels.

8. Stress-rupture test results at 1200 degrees Fahr. (650 degrees Cent.) are lower for annealed commercial 5 per cent chromium-molybdenum-titanium steels with titanium-carbon ratios of 5 to 8 than for 5 per cent chromium-molybdenum steels without titanium and with either 0.5 per cent maximum silicon or with 1 to 2 per cent silicon. But 5 per cent chromium-molybdenum-titanium steels with titanium-carbon ratios of 3.8 to 5.1 and silicon from 0.22 to 1.03 per cent, merely tempered at 1350 degrees Fahr. (730 degrees Cent.) instead of being annealed or normalized, have higher stress-rupture strength at 1200 degrees Fahr. (650 degrees Cent.) than the 5 per cent chromium-molybdenum steel without titanium. After 2000 hours under stress at 1200 degrees Fahr. (650 degrees Cent.), these titanium steels have excellent notch impact resistance.

9. The tensile, hardness, and impact properties of the 5 per cent

chromium-molybdenum-titanium steels merely tempered at 1350 degrees Fahr. (730 degrees Cent.) after hot working are satisfactory, so that annealing or normalizing is not required for them.

Acknowledgments

Numerous associates on the staff of the Titanium Alloy Manufacturing Company helped to obtain the data here presented and deserve a large share of the credit for whatever merit the work may possess. Special appreciation is due to A. F. Wolf, A. S. Yocco, and J. R. Lewis for making and testing the special steels, to H. W. Hudson of the National Tube Co. for supplying most of the commercial steels used, and to J. W. Freeman of the University of Michigan for making many of the high temperature rupture tests. The thanks of the author are also due to the Titanium Alloy Manufacturing Co. for providing the facilities required for the work, and for permission to publish the results.

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DISCUSSION

Written Discussion: By Evan F. Wilson, chief metallurgist, The Babcock & Wilcox Co., Barberton, Ohio.

The use of titanium additions to decrease the air hardening tendency of the 5 per cent chromium, 0.5 per cent molybdenum steel on cooling from above the critical range has been appreciated for some time. However, the quantitative effect of the titanium to carbon ratio on this hardening tendency has, previous to the presentation of this paper, been incompletely explored or understood. Mr. Comstock is therefore to be warmly congratulated for a most useful and well prepared contribution to our knowledge in this respect.

Titanium in steel contributes to the formation and stabilization of ferrite in temperature regions at which austenite normally forms, with a decrease in the proportion of the latter phase. This influence on the phase condition is similar to that displayed by chromium, molybdenum, silicon, aluminum, phosphorus and some other elements in sufficient amounts. By formation of titanium carbides, the contrary effect of carbon to promote austenite formation is also decreased. Even at quite elevated temperatures the low solubility of titanium carbides makes the carbon relatively unavailable for solution in the austenite constituent.

Because of these dual effects on the extent of austenite formation above the critical temperature, the titanium correspondingly modifies the state of agglomeration and distribution of precipitated carbides on transformation at subcritical temperatures. The obtainable mechanical properties of the steel are a reflection of this influence on the microstructure which is controlled by the phase proportions above the critical temperature.

The successful utilization of titanium in the 5 per cent chromium, molybdenum-bearing steels requires a more accurate control than has hitherto been commonly recognized. It necessitates the consideration of the additive ferrite-

forming tendency of the silicon and the opposite or austenite-favoring characteristics of manganese and carbon. By proper balance of these effective elements, the air hardenability, which is concerned with the relative amounts of austenite transformation products and the stabilized ferrite, can be carefully regulated. This regulation also requires that the chromium and molybdenum contents should not depart appreciably from their usual values for this alloy composition.

The optimum ratio of titanium to carbon for a specific application may not necessarily coincide with that favoring the minimum air hardening behavior. The most favorable impact properties are usually obtained where the ferrite stabilization effects are not too pronounced. Also, if the forging or other hot working of the steel is slight or negligible, control of this ratio should be such as to restrict excessive grain growth at elevated temperatures. Since the toughening effect from hot working is more important than that of partial recrystallization on normalizing of the titanium-bearing alloy, it is quite understandable that tempering only, rather than normalizing and tempering, is preferred for forged materials.

In view of the pronounced austenite-forming ability of nitrogen, a study of the nitrogen contents of the heats used in this investigation might be further contributory. It is possible that the effect of incidental nitrogen may be concerned in the greater scatter in hardness values attending the higher normalizing temperatures of Fig. 1. The unintentionally added aluminum content might also be significant.

The usefulness of titanium to reduce progressive scaling losses at high temperatures has been demonstrated by Mr. Comstock in previously reported tests. One of the presumptive reasons for this effect is the more dense and adherent scale favored by titanium oxide. A second contribution may be the increased availability of the solid solution chromium for oxidation resisting since the preferential formation of titanium carbide minimizes its tie-up as chromium carbides. Another benefit may be in the decreased diffusion of carbon from titanium carbides to the surface of the metal. There is consequently less tendency to form porous or flaky scale due to gaseous pressure from sublayer oxidation of carbon to the monoxide.

The stress-rupture test results reported by Mr. Comstock are consistent with other established observations that a moderate titanium addition raises the extrapolated stress value for rupture at the 10,000-hour period. This improvement may be presumed to be the result of increased carbide stability favored by the titanium during test exposure. However, the additive ferrite stabilization tendency from high titanium to carbon plus high silicon to manganese ratios can adversely influence the obtainable mechanical properties and reduce the rupture strength. The observed effect of heat treatment on rupture values is consistent with this influence of composition. Some discrepancies in rupture test results with titanium additions among different investigators may be explainable in consideration of the total composition of the steel tested.

Written Discussion: By Paul L. Daley, chief metallurgist, National Tube Co., Ellwood Works, Subsidiary U. S. Steel Corp., Ellwood City, Pa.

Mr. Comstock's paper has been extremely interesting to us since we furnished some of the steels tested and from time to time have had many interesting discussions with Mr. Comstock on the subject of the properties of titanium-

bearing chromium-molybdenum alloy steels. The paper presents information known in part to some, but to our knowledge this is the first time that results of so comprehensive a study have been presented.

Based on our experience with alloys of the types considered by Mr. Comstock, it is our opinion that the most significant feature in the addition of titanium to chromium-molybdenum alloy steel is often overlooked by many users of such steel. As shown by Mr. Comstock's test results, an effective amount of titanium reduces but does not eliminate air hardening. Mr. Comstock's data show, as we have experienced, that even with undesirably high Ti/C ratios some hardening may be experienced if the temperature reached has been sufficiently high. This is particularly significant in welding, since, no matter what the type of weld or the filler metal used, a hardened zone will be found. As this applies to columbium-bearing steel as well, it is useful to know that tempering will readily reduce the hardness and improve the notched-bar strength. By the preheat and postheat practice employed for nontitanium-bearing 5 per cent chromium, 0.5 per cent molybdenum steel our experience indicates that the hardened zone can be substantially eliminated, or at least controlled to a maximum hardness under 250 Brinell hardness number (usually under 207 Brinell hardness number), whereas without this practice the hardness has been found to be as high as 321 Brinell hardness number. One point not generally considered in connection with the welding of these steels with nonhardening austenitic filler metal is the fact that, because of dilution of the filler metal with base metal, a new alloy composition results, at least at the interface. When using 18 per cent chromium-8 per cent nickel filler metal it has been shown that, because of this dilution, a small amount of alloy martensite may be found at the base of the weld, or more commonly, along the sides of the weld. Although such a condition is not always found and as tests and experience have not proved it to be harmful, it now is our opinion that the use of 25 per cent chromium-12 per cent nickel or, better still, 25 per cent chromium-20 per cent nickel filler is to be preferred, because dilution is far less likely to produce a martensitic alloy in the weld. Some objection to the use of austenitic filler metal has been raised at times, but we have found that, in arc welding, austenitic steel electrodes manipulate more readily than do the columbium-bearing 5 per cent chromium, 0.5 per cent molybdenum electrodes. Consequently, it is easier to produce sound welds with the austenitic steel electrodes.

While the titanium-bearing 5 per cent chromium-0.5 per cent molybdenum alloy nominally falls in the pearlitic steel classification, actually, as indicated in Mr. Comstock's paper in his description of microstructures, the steel is more nearly like a ferritic alloy in its behavior. While the A_1 temperature on heating is at about 1700 degrees Fahr. (925 degrees Cent.), in an alloy containing a Ti/C ratio of 4/1 minimum, no A_2 point is exhibited up to the melting temperature. For this reason it is probably not correct to refer to "normalizing" in connection with this alloy. This would account for the need to continue hot working to low temperatures to produce a fine structural grain size and the inability to refine the grain size by heat treatment alone. Experience with the high silicon, titanium-bearing 5 per cent chromium, 0.5 per cent molybdenum grade has been limited but unfortunate. While this alloy possesses very desirable oxidation resistant properties, difficulty with grain coarsening, lack of toughness and cold workability has been experienced. Since silicon is a ferrite former, too, the grain coarsening

troubles could be expected, but the lack of toughness and cold workability does not appear to be entirely related to grain size. A similar condition has been found to exist in 7 and 9 per cent chromium alloys with 0.5 and 1 per cent molybdenum and is being investigated. A remedy has apparently been found, but no satisfactory explanation as to its cause has yet been reached.

Written Discussion: By C. L. Clark, research metallurgical engineer, The Timken Roller Bearing Co., Steel and Tube Division, Canton, Ohio.

The producers and users of high temperature steels should be indebted to Mr. Comstock for the data as to the effect of variations in chemistry and heat treatment on the properties of the 5 per cent chromium-molybdenum type steel. In general, it is gratifying to find that the standard analysis which has been developed over a period of years agrees fairly well with that which would be indicated by Mr. Comstock's work.

We believe Mr. Comstock to be in error when he states that current specifications place no limit on the maximum amount of titanium permissible. The two specifications to which this analysis is generally produced, A.S.T.M. A158-44T and A.S.T.M. A213-44, both place an upper limit of 0.70 per cent on the titanium content and a lower limit of 4 times the carbon content. In both specifications the carbon is specified as 0.12 maximum and most commercial heats will be in the range of 0.09 to 0.11 per cent carbon.

It must be further recognized that because of the high affinity of titanium for oxygen, and perhaps nitrogen as well, its recovery, when added to molten steel, is variable and consequently it would be impossible to melt to too close a titanium to carbon ratio.

It is generally agreed that increased silicon content greatly improves the oxidation resistance of the 5 per cent chromium-molybdenum type steels and the only reason why an increased silicon content has not been added to the titanium modification is that its main commercial applications to date have been confined to temperatures at which oxidation resistance is not of primary importance. In this same connection, our own work has not shown the oxidation resistance to be improved by the addition of titanium. Further, it is our belief that the differences in the oxidation behavior of Steels 17 and 19, of Table IX, especially at 1100 and 1300 degrees Fahr. (595 and 705 degrees Cent.), are within the experimental error.

We agree with Mr. Comstock that the 5 per cent chromium-molybdenum-titanium steel should be normalized or tempered, rather than annealed. However, if cold drawing is involved during processing, it is believed the tempering temperature should be higher than 1350 degrees Fahr. (705 degrees Cent.), in order to obtain a proper yield-to-tensile ratio.

Oral Discussion

H. A. Smith:[2] I should like to ask Mr. Comstock a question about the application of some of these steels in the hardened and tempered condition. In the aircraft industry, we are now assembling with CAA approval landing gear structures by means of silver brazing. We braze heat treated alloy steel parts and want to retain as much hardness as possible after brazing. This means we must have a steel with a low rate of tempering. We would like to maintain a minimum

[2] Chief metallurgist, Beech Aircraft Corp., Wichita, Kansas.

Rockwell of C-35 after induction brazing in ¼ to ⅜-inch sections. This means in a period of about 10 minutes the steel is heated from ambient temperature to 1250 to 1300 degrees Fahr. (675 to 705 degrees Cent.). The steel then cools normally. In the ordinary constructional type steels, such as the 8600, 8700, and 4300 series, about all we can get is C-30 after brazing. I am wondering whether or not a steel similar to that specified as Number 2 with normal manganese and silicon and with a low titanium-carbon ratio will attain such a hardness and possess such a tempering rate that after brazing the Rockwell hardness would be C-35 or above.

Author's Reply

The interesting comments submitted as discussion of this paper constitute a valuable addition to the data recorded, and are much appreciated by the author. Mr. Wilson's contribution especially indicates a thoughtful study of the paper and its implications, and provides some fundamental explanations of the results. The paper was undoubtedly deficient in failing to include an adequate discussion of these features, and Mr. Wilson should be thanked for supplying it. Although the effect of nitrogen is discussed briefly on page 93, it is admitted that the nitrogen contents of all the steels would have been of interest. It is extremely unlikely, however, that any of those reported in the paper elsewhere than in Table VI contained enough nitrogen to affect the results as much as is indicated in that table.

Mr. Daley's comments are also valuable, especially in regard to welding. It is noted that difficulty has been experienced in several respects with the high silicon grade of 5 per cent chromium-molybdenum-titanium steel. According to our data, as reported in the paper, many of these difficulties might be overcome by raising the manganese content of this grade to between 0.75 and 1.00 per cent, and restricting the titanium-carbon ratio to between 3.5 and 5.5. It would be very interesting if the National Tube Co. could check our findings in this respect.

Dr. Clark has misunderstood the paper in referring to a statement that "current specifications place no limit on the maximum amount of titanium permissible" in these steels. On page 107, lines 15 and 16, it is stated that the specifications should set a maximum limit on the titanium-carbon ratio, and not merely on the titanium content. Since there is no *minimum* limit specified for carbon, the titanium-carbon ratios permitted under the present specifications may be anything between 4 and 12 or more, and this variation is entirely too great to insure reasonably consistent mechanical properties. Commercial steels submitted to us as representing usual practice had titanium-carbon ratios between 5 and 8, and most of them had lower carbon than the 0.09 per cent reported by Dr. Clark. With carbon above that value, these steels contain too much titanium carbide, and we believe that carbon around 0.05 or 0.07 per cent is much better. The titanium, however, should also be held lower, so that the titanium-carbon ratio does not exceed 5 or 6. With proper care in deoxidizing both steel and slag with aluminum before adding titanium to the steel, it should not be too difficult to secure the necessary consistency in recoveries.

We are sorry to note that our experience with oxidation tests and the effect of heat treatments is at variance with Dr. Clark's. It is admitted that the effect of titanium on the weight losses of these steels due to oxidation is slight, but the

prevention of pitting such as occurred in samples 21 and 22, Fig. 13, seems rather important. In regard to the yield ratio, our tests as reported in Table XIII indicate that tempering at 1350 degrees Fahr. (730 degrees Cent.) gave values of 66 to 92 per cent, which are higher than those obtained on heating at higher temperatures, such as are recommended by Dr. Clark for cold-drawn shapes. It is not clear, at least to the author, why cold-drawn 5 per cent chromium-molybdenum-titanium steels should need to be heated at appreciably higher temperatures in practical processing, since it has now been pointed out that better properties generally are obtained by avoiding such treatment.

In reply to Mr. Smith, there seems to be no reason why, if he is interested in increased hardness, he should contemplate using a titanium steel. Possibly the ordinary 5 per cent chromium, 0.5 per cent molybdenum steel, with 0.10 to 0.15 per cent carbon, might resist tempering to the extent that he desires. The hardness values of the titanium steels in Table XIII, which were tempered at 1350 degrees Fahr. (730 degrees Cent.), would indicate that none of them would be as hard as Rockwell C-35, or 330 Brinell, under the conditions described by Mr. Smith.

IRON-MANGANESE ALLOYS

The Properties of Cold-Worked and Heat Treated Alloys Containing 1 to 7 Per Cent Manganese

By R. S. Dean, J. R. Long, T. R. Graham and R. G. Feustel

Abstract

Iron-manganese alloys containing 1 to 7 per cent manganese have been shown to be amenable to cold working. Properties of material reduced 20, 40, 60, and 80 per cent in thickness by cold rolling are reported. In the soft condition the tensile strength is increased by about 11,000 pounds per square inch for each 1 per cent manganese. The elongation drops rapidly at first (5 per cent for each 1 per cent manganese) and then more slowly (1 per cent for each 1 per cent manganese) at 7 per cent manganese. Cold working to 80 per cent reduction produces a tensile strength of 192,600 pounds per square inch, an elongation of 2.7 per cent in 2 inches, and a hardness of Rockwell C-38. Normalizing and annealing of cold-worked material produces essentially the same properties for a given temperature of treatment. Low-manganese alloys show the usual softening with increasing annealing temperature. High-manganese alloys show a secondary hardening at annealing temperatures above 1100 degrees Fahr. (595 degrees Cent.). Intermediate alloys show this same hardening with a resoftening above 1400 degrees Fahr. (760 degrees Cent.). Tempering of normalized material produces minimum strength and hardness independent of normalizing temperatures.

DATA available in the literature on the mechanical properties of iron-manganese alloys have been largely restricted to forged, or forged and heat treated material and to the general effects of carbon content. Individual investigators (1), (2), (3)[1] have reported

[1]The figures appearing in parentheses pertain to the references appended to this paper.

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A paper presented before the Twenty-seventh Annual Convention of the Society held in Cleveland, February 4 to 8, 1946. Of the authors, R. S. Dean is assistant director, Bureau of Mines, U. S. Department of the Interior, Washington, D. C., J. R. Long and T. R. Graham are senior metallurgists, and R. G. Feustel is assistant metallurgist, Bureau of Mines, Intermountain Experiment Station, Salt Lake City, Utah. Manuscript received June 8, 1945.

high ductility for the alloys, which would indicate a large capacity for plastic deformation, but no systematic data on this point have been published. It, therefore, appeared desirable to examine these alloys to determine their capacity for cold working and the effect of this deformation and subsequent heat treatments on their physical properties. The work of the U. S. Bureau of Mines on alloys made with electrolytic manganese has, accordingly, included a study of iron-manganese alloys, with particular attention to their properties in the cold-worked state. While the project covers a broad range of compositions, only those up to 7 per cent manganese are considered in this paper. The results on alloys with higher manganese contents will be presented at a later date.

Alloy Preparation and Fabrication

The alloys for this work were prepared from Armco ingot iron and electrolytic manganese cathode chips. They were melted in a high-frequency induction furnace in 50-pound heats and chill-cast into tapered ingots 3.5 by 3.5 by 12 inches with hot-tops to reduce the pipe. The alloying was readily accomplished by adding the cathode chips to the molten iron. No deoxidation other than that obtained by the manganese was considered necessary. The chill-cast ingots produced were sound, had excellent surfaces, and were reasonably clean. Chemical analyses of the alloys are given in Table I. While no special efforts were made to produce alloys of the highest purity, the carbon, silicon, phosphorus, and sulphur contents are reasonably low and the alloys may be considered as simple binary alloys. For simplicity of reference to composition, the alloys shall be referred to throughout the rest of the paper as 1 per cent for the 1.15 per cent manganese, 2 per cent for the 1.85 per cent manganese, and so on to 7 per cent for the 6.95 per cent manganese.

Table I
Chemical Analysis of Materials Investigated, Per Cent

Alloy No.	Manganese	Carbon	Sulphur	Phosphorus	Silicon
Armco	0.029	0.016	0.02	0.002	0.005
1	1.15	0.02	0.03	0.019	0.01
2	1.88	0.02	0.03	0.009	0.01
3	3.15	0.02	0.03	0.004	0.01
4	4.12	0.02	0.03	0.007	0.01
5	5.10	0.02	0.03	0.005	0.01
6	5.85	0.01	0.03	0.003	0.01
7	6.55	0.01	0.03	0.007	0.01
8	6.95	0.01	0.03	0.003	0.01

Forging was carried out with an initial temperature of 2000 degrees Fahr. (1040 degrees Cent.) and a finishing temperature of 1600 degrees Fahr. (871 degrees Cent.). The scale, while heavy, was not excessive and was readily removed by the first few hammer blows. The alloys forged well and were finished into 1 by 4-inch slabs. Following this they were hot-rolled into $\frac{5}{8}$ by 4-inch plate at 1600 degrees Fahr. (870 degrees Cent.) with frequent reheating to avoid rolling below 1450 degrees Fahr. (785 degrees Cent.). Although this temperature range was chosen arbitrarily, it proved very satisfactory. Rolled-in scale and other surface defects were removed by machining after air cooling from the rolling temperatures. All of the plates were next given a 35 to 40 per cent reduction by cold rolling to bring them to a thickness of 0.312 inch, the starting point for the rolling schedule calculated to produce $\frac{1}{16}$-inch sheet finished with definite amounts of cold deformation. The Rockwell hardness was determined after every 0.040 to 0.050-inch reduction to follow the work hardening. However, hardness was not found to be a good criterion of the effect of plastic deformation in this first cold rolling. This was particularly true of those above 3 per cent manganese. The hardness actually decreased with the initial cold working and in some cases this decrease persisted beyond 30 per cent reduction in thickness. The drop in hardness is considered to be due to decomposition of the martensitic constituent formed by the transformation of austenite during air cooling (4).

Annealing the cold-rolled material for further processing required considerable experimentation. Alloys containing 1 and 2 per cent manganese could be readily softened by heating to 1450 degrees Fahr. (785 degrees Cent.) for 30 minutes and air cooling. The higher manganese alloys, however, ranged from Rockwell C-16 to C-33 after rolling and were unaffected by the above treatment. Variation in cooling rate and in the higher annealing temperature produced but little effect, no doubt due to the formation of martensitic phases in the alloy. Tempering treatments decomposed this martensite and produced a reasonable degree of softening. A procedure of heating to 1450 degrees Fahr. (785 degrees Cent.) for 30 minutes followed by reheating to 1150 degrees Fahr. (620 degrees Cent.) for 1 hour and air cooling was finally chosen since it produced the minimum hardness. This treatment was then used for all intermediate annealing of alloys containing 3 and more per cent manganese.

The drop in hardness with initial deformation amounts to only

a few points with material annealed in this way and occurs only in the first few per cent reduction; before 10 per cent reduction has been reached the hardness again increases in a normal fashion. In the absence of anomalies produced by the decomposition of the martensitic constituent, hardness may be readily used to follow the strain-hardening produced by cold work.

The alloys were finished into $\frac{1}{16}$-inch thick sheet by rolling and annealing schedules calculated to finish up with 20, 40, 60, and 80 per cent reduction by cold rolling. The major portion of each alloy was finished at 60 per cent reduction to provide material for determining the effect of heat treatment at various temperatures. The annealed or zero cold-worked state was obtained by annealing material that had been finished by 60 per cent cold reduction using the treatments described above.

The 1 and 2 per cent manganese alloys, as well as the Armco iron, were cold-rolled by 0.025 to 0.035-inch reduction per pass to within a few thousandths of the finished size, then by 0.002 to 0.004 inch per pass to the finished 0.0625-inch sheet. All three compositions were comparable in their rolling behavior, although the 2 per cent manganese alloy did offer slightly more resistance to deformation. The alloys containing 3 to 5 per cent manganese were similarly processed by reductions of 0.015 to 0.025 inch per pass. Alloys containing greater than 5 per cent manganese were more carefully handled due to the higher yield strength and hardness levels and were reduced by 0.003 to 0.005 inch per pass through the entire range.

Annealed Properties

Data on the mechanical properties as affected by cold rolling are given in Table II, arranged as a function of per cent reduction for each alloy. The values given are averages obtained from four specimens tested in each condition. These data are also plotted in the graphs as a function of manganese content. The effect of manganese on tensile strength is shown in Fig. 1, yield strength (0.1 per cent offset) in Fig. 2, elongation in Fig. 3, and hardness in Fig. 4.

In the annealed condition the first 1 per cent manganese has but little effect, the tensile strength is slightly higher, the yield strength is slightly lower, the elongation is about 5 per cent lower, and the hardness is no higher than that of the Armco iron. Above this, the strength and hardness increase regularly while the elongation decreases regu-

larly with increasing manganese content. The tensile strength increases at a rate of about 11,000 pounds for each 1 per cent manganese, reaching 115,000 pounds per square inch at 7 per cent. The yield strength for 0.1 per cent offset appears to be somewhat high

Table II
Variation in Physical Properties with Cold Working

Alloy	Reduction in Thickness Per Cent	Tensile Strength, Lbs. Per Square Inch	Yield Strength 0.1 Per Cent Offset, Lbs. Per Square Inch	Proportional Limit, Lbs. Per Square Inch	Modulus $E \times 10^{-6}$	Rockwell Hardness	Elongation Per Cent
Armco	0*	43,700	36,300	.....	29.1	55B	39.6
ingot	20	72,200	67,500	59,500	25.9	87B	9.5
iron	40	84,500	81,800	72,850	28.8	9C	8.3
	60	96,100	97,500	89,200	30.5	16C	3.3
	80	111,200	.....	96,800	30.9	19C	3.1
1.15% Mn	0*	48,000	32.000	.....	24.7	55B	34.7
	20	72,360	69,700	63,100	26.7	4C	6.7
	40	90,800	90,300	82,100	29.2	13C	3.1
	60	102,800	102,600	94,100	26.3	17C	3.1
	80	110,900	107,900	96,400	27.2	19C	2.3
1.88% Mn	0*	58,800	28,000	22,400	26.3	63B	29.3
	20	76,000	71,100	55,200	23.8	7C	7.9
	40	95,100	90,300	74,900	23.7	15C	5.2
	60	108,900	102,200	80,400	23.1	19C	4.7
	80	118,200	109,000	83,000	23.0	21C	4.6
3.15% Mn	0*	72,000	37,300	27,800	27.7	80B	25.0
	20	87,900	82,200	64,800	23.4	12C	6.7
	40	100,500	93,500	71,400	22.9	18C	5.1
	60	111,900	102,100	76,900	25.6	20C	4.7
	80	128,000	114,300	79,600	23.8	26C	4.5
4.12% Mn	0*	85,100	55,500	44,100	25.7	7C	20.4
	20	102,600	96,000	73,600	25.0	18C	6.8
	40	117,500	108,100	85,500	21.4	24C	4.5
	60	124,800	113,700	84,900	24.5	26C	4.4
	80	153,100	136,600	101,300	25.6	31C	4.3
5.10% Mn	0*	94,000	61,300	48,600	27.6	13C	20.1
	20	106,700	99,200	77,400	24.1	21C	7.4
	40	128,000	114,000	84,300	25.5	27C	5.5
	60	136,100	122,500	93,000	25.6	28C	5.6
	80	164,800	.....	104,200	25.9	34C	3.4
5.85% Mn	0*	97,800	65,800	50,900	25.8	15C	19.3
	20	117,200	107,700	83,000	23.9	25C	7.1
	40	129,400	117,700	89,000	24.6	28C	4.7
	60	149,500	134,100	102,700	24.8	32C	3.9
	80	173,900	153,600	112,900	25.1	35C	3.7
6.55% Mn	0*	107,500	71,300	54,500	25.7	19C	18.6
	20	132,000	118,900	91,300	26.1	30C	7.5
	40	156,000	141,000	108,900	25.5	33C	4.2
	60	171,500	155,500	122,000	23.2	34C	4.3
	80	187,200	170,500	152,200	25.4	37C	3.6
6.95% Mn	0*	114,900	70,000	56,000	25.5	23C	16.2
	20	148,900	135,000	101,500	25.4	32C	3.8
	40	165,600	146,000	111,700	25.5	35C	3.6
	60	178,400	159,300	115,200	26.1	37C	3.5
	80	192,500	160,500	121,200	24.3	38C	2.7

*See text for annealed condition or 0 per cent cold work.

for the Armco iron. Disregarding the high yield strength for Armco iron and projecting roughly from 1 to 7 per cent, manganese gives an increase in yield strength of approximately 6000 pounds for each 1 per cent manganese. The elongation in 2 inches drops at an initial

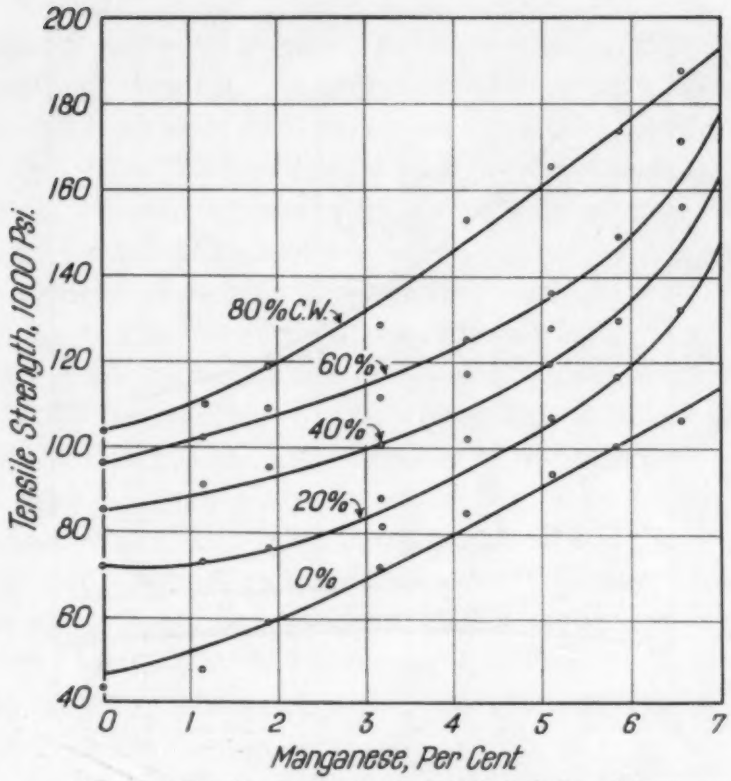


Fig. 1—Tensile Strength of Cold-Rolled Iron-Manganese Alloys Plotted as a Function of Manganese Content for 0, 20, 40, 60 and 80 Per Cent Reduction in Thickness.

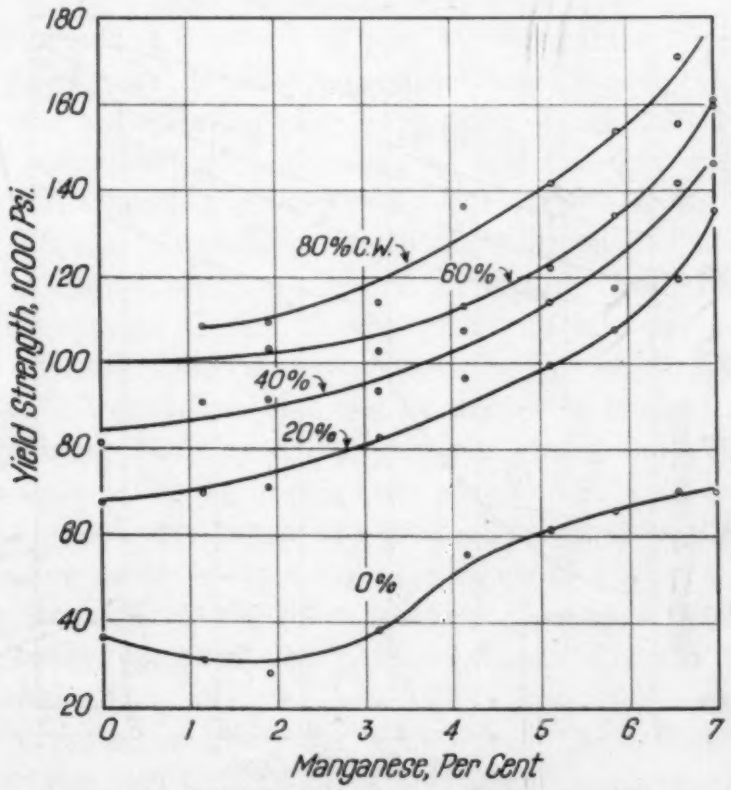


Fig. 2—Yield Strength of Cold-Rolled Iron-Manganese Alloys Plotted as a Function of Manganese Content for 0, 20, 40, 60 and 80 Per Cent Reduction in Thickness.

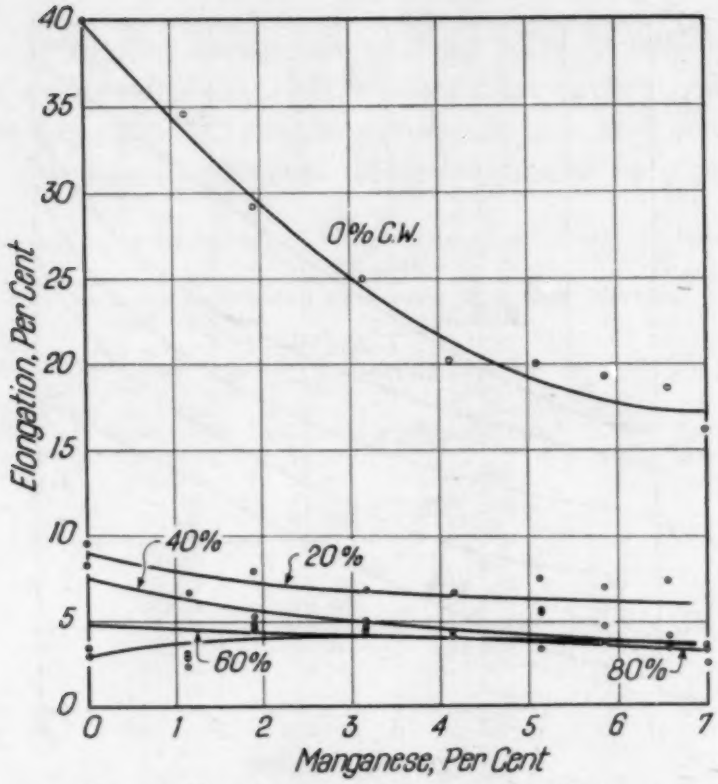


Fig. 3—Elongation of Cold-Rolled Iron-Manganese Alloys Plotted as a Function of Manganese Content for 0, 20, 40, 60 and 80 Per Cent Reduction in Thickness.

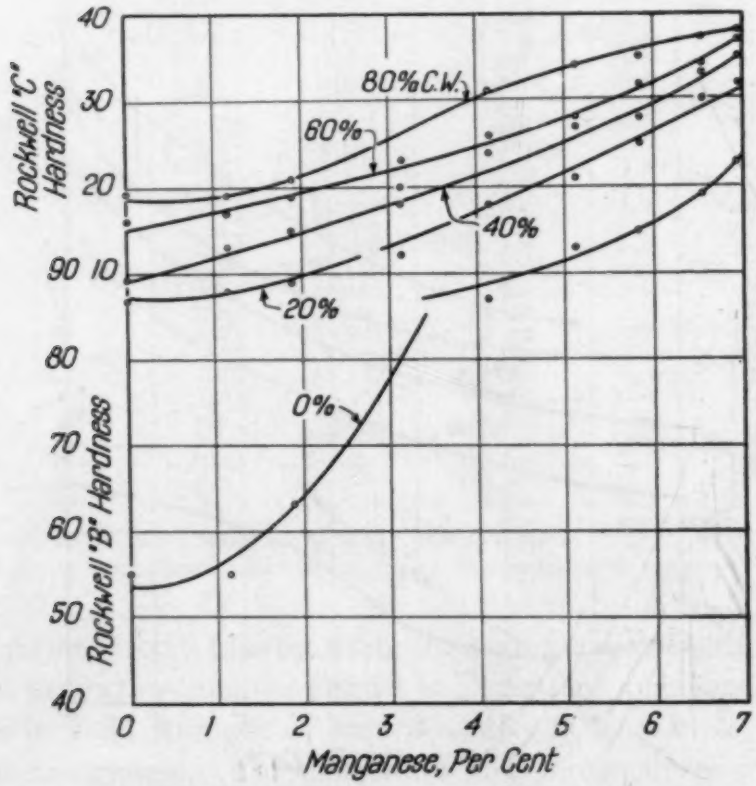


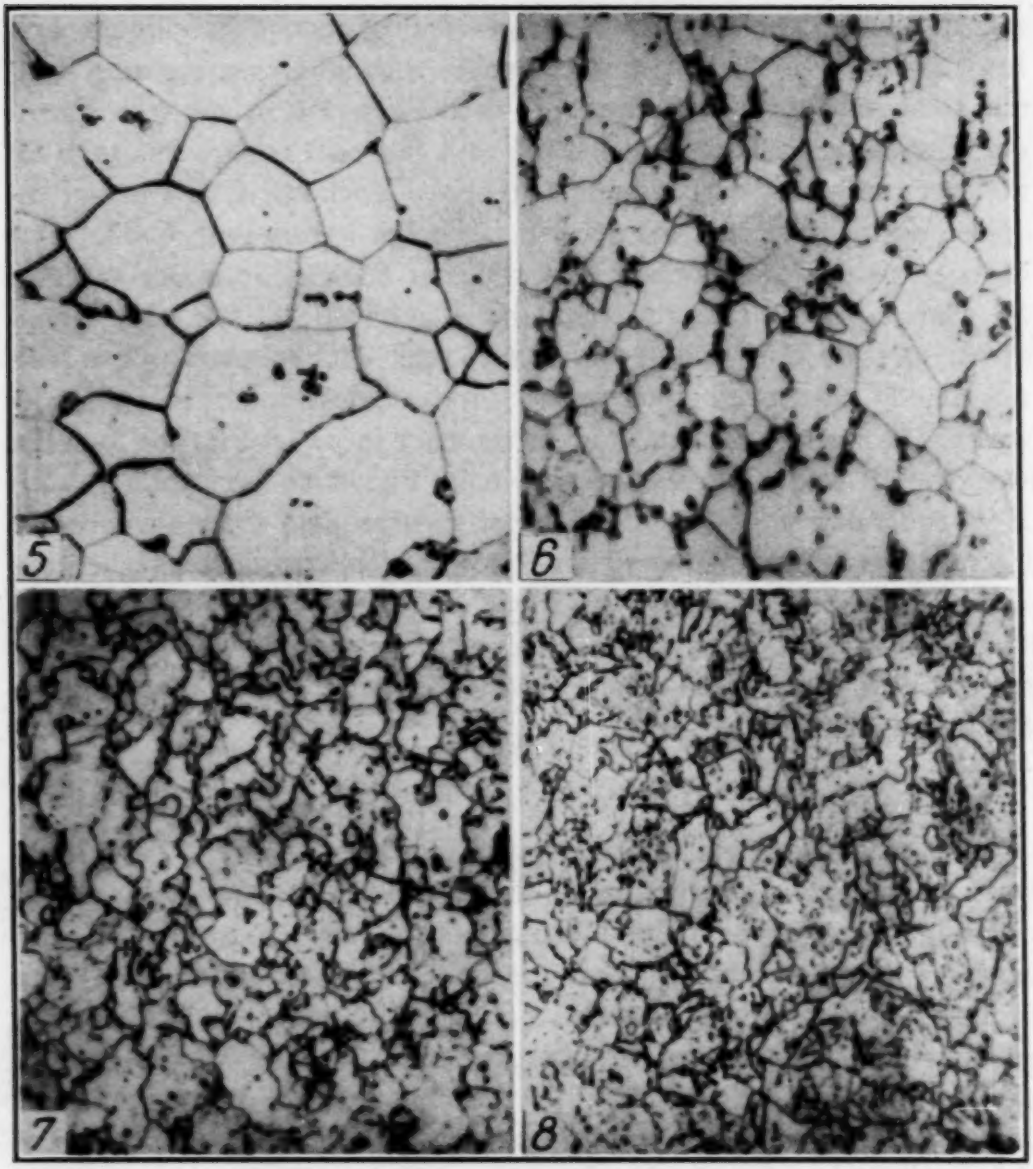
Fig. 4—Rockwell Hardness of Cold-Rolled Iron-Manganese Alloys Plotted as a Function of Manganese Content for 0, 20 40, 60 and 80 Per Cent Reduction in Thickness.

rate of about 5 per cent for each 1 per cent of manganese for the first few per cent and then the rate decreases with an increasing alloy content to about a 1 per cent drop with each 1 per cent manganese. Hardness increases from Rockwell B-55 at 1 per cent manganese to B-90 at 4 per cent. This corresponds roughly to Rockwell C-7. Further increase up to 7 per cent manganese raises the hardness to C-23. The change in slope of the curve at 4 per cent manganese is due more to the change in scale than to irregularities in hardness. The wide spread of hardness encountered in these alloys makes it necessary to use both the B and C Rockwell scales to avoid meaningless values below Rockwell C-10 or above B-100. For convenience in presenting the data the Rockwell B and C scales are shown in Fig. 4 as overlapping, with B-90 equivalent to C-10. This is not accurate and the scales are not interchangeable over an extended range. However, cross checking with these alloys has shown Rockwell B-100 to be equivalent to C-20 and B-90 to C-8. To avoid confusion in the use of these two scales, the curves are broken to separate the data plotted on the B scale from those plotted on the C scale.

Microstructures

Typical microstructures are given in Figs. 5 to 12. They illustrate the changes in structure of the iron-manganese alloys with increasing manganese. These structures are representative of the zero per cent cold-worked and the ready-to-finish condition of the individual compositions investigated. The 2 per cent alloy shows a small amount of a martensitic constituent formed from the gamma phase on cooling. The amount of this constituent increases with increasing manganese and with simple air cooling from 1450 degrees Fahr. (785 degrees Cent.) should constitute 100 per cent of the structure at about 5 per cent manganese. The constitution and appearance of this phase are modified by tempering treatments, and the alloys containing 3 and more per cent manganese show this in Figs. 7 to 12 inclusive. The high-manganese alloys of Figs. 11 and 12 show an acicular type of structure which gradually changes to a particle type of distribution in the lower-manganese alloys of Figs. 7 and 8. The transition from one to the other is nicely illustrated by Figs. 9 and 10. The tempering results in increased amounts of the stable alpha solid solution.

The variation in grain size with manganese content shown by these structures is of little significance, since the grain size here is con-

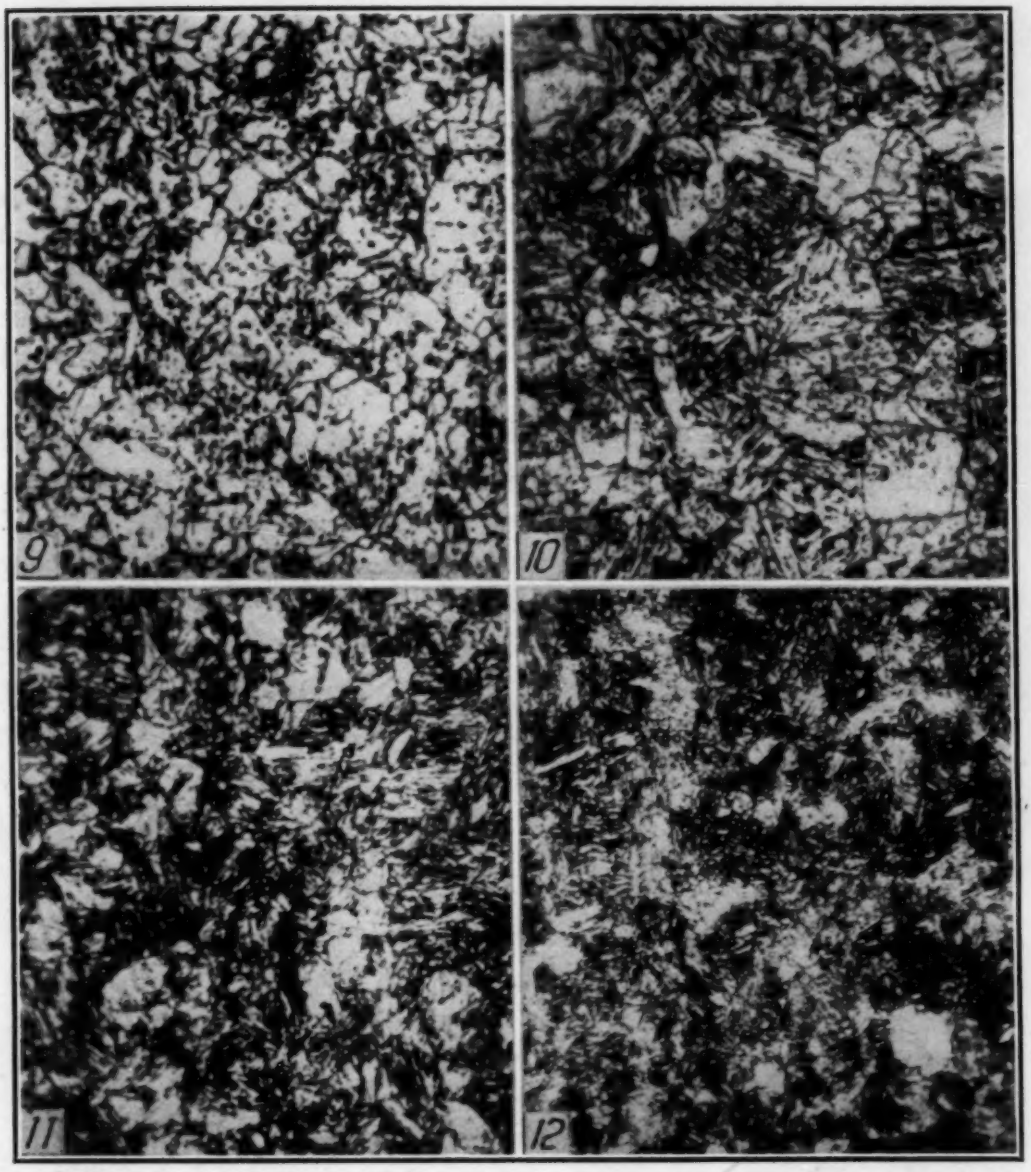


Figs. 5-8—Typical Structures of Iron-Manganese Alloys After Heat Treatment. × 500. Manganese content of each specimen is as follows: Fig. 5—1.15 per cent. Fig. 6—1.88 per cent. Fig. 7—3.15 per cent. Fig. 8—4.12 per cent. Figs. 5 and 6—normalized from 1450 degrees Fahr. (785 degrees Cent.). Figs. 7 and 8 normalized from 1450 degrees Fahr. (785 degrees Cent.) and tempered 1 hour at 1150 degrees Fahr. (620 degrees Cent.).

ditioned primarily by the alpha to gamma transformation temperature. As they were all heated to the same temperature some were well above and some below their respective transformation temperatures.

Cold-Worked Properties

In the cold-worked state the strength and hardness increase regularly and are still rising after 80 per cent reduction in thickness by



Figs. 9-12—Typical Structures of Iron-Manganese Alloys, Cold-Worked, Normalized from 1450 Degrees Fahr. (785 Degrees Cent.) and Tempered for 1 Hour at 1150 Degrees Fahr. (620 Degrees Cent.). × 500. Manganese content of each specimen is as follows: Fig. 9—5.10 per cent. Fig. 10—5.85 per cent. Fig. 11—6.55 per cent. Fig. 12—6.95 per cent.

cold rolling. The increase of tensile strength with manganese content for 80 per cent cold working is slightly greater than that for the annealed condition. It amounts to about 13,000 pounds per square inch for each 1 per cent manganese. The other curves of Fig. 1 are roughly parallel, indicating about the same rate for most of the alloys, although they are somewhat steeper at the high-manganese end. The work hardening rates for all of the alloys are about the same for the range of manganese content given. The 7 per cent alloy has a tensile

strength of 192,000 pounds per square inch after 80 per cent cold work, which is almost double its annealed strength. The other alloys have increased in about the same proportion. These strength values are not unusually high but are significant in possible uses of these alloys and in their fabrication by cold working.

The elongation drops rapidly with the first 20 per cent reduction and more slowly for greater reduction showing little difference for 60 and 80 per cent reduction. Armco iron and the 1 per cent manganese alloy show less elongation after 60 to 80 per cent reduction than the 2 to 7 per cent alloys. Two or more per cent manganese is apparently effective in increasing the ductility of the more severely cold-worked metal and the increase is significant at least through 7 per cent manganese. The percentage elongation for a moderate amount of cold work is also independent of the manganese content beyond 2 per cent and is quite insensitive to both cold work and manganese content beyond 40 per cent reduction. It is, therefore, obvious that in iron-manganese alloys cold-worked to the extent of 20 per cent or more, higher strengths may be obtained by higher manganese contents without sacrifice of ductility.

Properties of Heat Treated Material

Test specimens cold-worked to 60 per cent reduction were heated for 8 hours in a protective atmosphere of helium at temperatures ranging from 800 to 1700 degrees Fahr. (425 to 925 degrees Cent.). One set of these was cooled in the furnace, and two sets were air-cooled. One of the air-cooled sets was then given a tempering treatment of 1 hour at 1150 degrees Fahr. (620 degrees Cent.) and again air-cooled. This temperature was selected because most of the alloys showed minimum hardness and tensile strength with maximum elongation between 1100 and 1200 degrees Fahr. (595 and 650 degrees Cent.). It appeared to be high enough to alter the martensite producing a maximum amount of the alpha solid solution. The properties obtained on heat-treated specimens are given in Tables III, IV, V, and VI.

These treatments show considerable variation in the properties with variation in annealing temperature. It will be noted also that the properties are not materially affected by cooling rates but are greatly influenced by tempering at 1150 degrees Fahr. (620 degrees Cent.). Analysis of the data for the furnace-cooled specimens show

Table III

Tensile Strength of Cold-Worked and Heat Treated Iron-Manganese Alloys

Alloy	60% C. W.	Specimens heated 8 hours at indicated temperature and furnace-cooled — Degrees Fahr. 800	900	1000	1100	1200	1300	1400	1500	1600	1700
Armco	96,100	73,500	.....	.....	41,000	.....	.....	.....	.....	.....	.....
1.15% Mn	102,800	90,400	83,400	70,900	49,500	45,700	43,600	44,300	45,100	46,900	47,100
1.88% Mn	108,900	106,700	92,700	62,500	54,800	50,900	48,500	54,900	55,500	52,400	50,900
3.15% Mn	111,900	106,500	96,400	70,000	62,600	65,100	90,900	91,800	75,500	69,300	66,000
4.12% Mn	124,800	115,100	110,200	90,300	82,700	89,500	115,000	99,900	96,600	92,700	93,400
5.10% Mn	136,100	124,100	121,200	101,000	94,500	108,000	123,400	111,400	109,000	105,700	107,900
5.85% Mn	149,500	136,800	121,100	108,000	102,300	119,100	126,900	117,400	115,100	113,200	106,600
6.55% Mn	171,500	139,700	125,900	114,400	108,900	134,300	132,000	127,900	125,300	125,200	124,100
6.95% Mn	178,400	148,100	130,600	118,200	113,200	137,000	131,900	130,900	133,000	129,300	130,100
		Specimens heated 8 hours at indicated temperature and air-cooled									
Armco	96,100	75,500	.....	.....	46,300	.....	.....	.....	.....	.....	.....
1.15% Mn	102,800	91,800	87,400	81,000	50,500	50,300	43,500	44,700	47,200	55,000	51,300
1.88% Mn	108,900	106,200	99,300	89,000	59,000	54,700	50,400	62,500	59,200	58,100	56,000
3.15% Mn	111,900	106,800	103,300	93,700	63,200	68,300	84,200	94,200	85,400	81,900	77,600
4.12% Mn	124,800	117,100	114,600	104,300	82,300	94,100	108,600	104,700	107,200	100,900	94,300
5.10% Mn	136,100	124,600	124,400	113,000	93,600	110,700	122,100	121,200	120,100	108,800	115,000
5.85% Mn	149,500	137,600	126,900	113,800	99,300	115,400	129,900	128,700	121,100	121,900	126,000
6.55% Mn	171,500	140,500	132,400	119,300	103,900	127,500	.....	134,000	135,100	130,600	131,500
6.95% Mn	178,400	147,800	136,300	124,400	112,500	132,900	137,700	136,500	130,200	132,100	134,500
		Specimens heated 8 hours at indicated temperature, air-cooled, and reheated 1 hour at 1150° F.									
Armco	96,100	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....
1.15% Mn	102,800	.....	.....	.....	.....	47,800	45,500	43,100	46,800	50,900	51,600
1.88% Mn	108,900	.....	.....	.....	.....	52,200	50,200	52,500	56,200	55,900	54,500
3.15% Mn	111,900	.....	.....	.....	.....	67,000	70,900	67,700	70,800	69,600	67,100
4.12% Mn	124,800	.....	.....	.....	.....	89,700	85,400	80,100	79,200	77,700	73,800
5.10% Mn	136,100	.....	.....	.....	.....	100,400	93,500	86,900	84,600	80,700	79,900
5.85% Mn	149,500	.....	.....	.....	.....	103,700	95,300	91,800	88,500	85,900	88,300
6.55% Mn	171,500	.....	.....	.....	.....	.....	.....	102,600	98,600	95,400	97,300
6.95% Mn	178,400	.....	.....	.....	.....	118,800	104,800	106,100	103,200	100,500	101,100

Table IV
Proportional Limit of Cold-Worked and Heat Treated Iron-Manganese Alloys

Alloy	60% C. W.	Specimens heated 8 hours at indicated temperature and furnace-cooled — Degrees Fahr. 800	900	1000	1100	1200	1300	1400	1500	1600	1700
Armco	89,200	70,700	.....	.....	38,600	.....	.....	.....	.....	.....	.....
1.15% Mn	94,100	79,800	77,100	65,900	37,100	34,900	29,300	24,100	27,900	31,400	29,900
1.88% Mn	80,400	96,200	92,900	62,500	48,900	43,900	36,200	16,200	34,300	30,100	28,600
3.15% Mn	76,900	95,900	92,100	59,700	49,600	46,500	29,400	39,400	28,200	25,800	27,100
4.12% Mn	84,900	104,500	104,500	89,400	84,200	64,800	42,700	48,400	52,000	44,100	44,400
5.10% Mn	93,000	114,000	118,700	101,000	93,000	49,600	48,400	54,000	49,800	51,300	50,300
5.85% Mn	102,700	120,500	116,000	103,800	92,500	31,900	38,300	55,000	56,500	48,600	48,200
6.55% Mn	122,000	130,100	119,900	106,300	105,300	41,600	50,600	59,500	57,600	47,300	56,500
6.95% Mn	115,200	137,400	126,800	105,800	101,800	39,500	47,100	66,200	61,400	48,700	57,300
		Specimens heated 8 hours at indicated temperature and air-cooled									
Armco	89,200	71,900	.....	.....	47,400	.....	.....	.....	.....	.....	.....
1.15% Mn	94,100	79,500	79,200	76,000	48,000	48,400	29,000	20,400	29,100	44,600	39,100
1.88% Mn	80,400	94,700	89,900	88,500	58,400	49,600	34,600	17,200	19,800	19,600	27,700
3.15% Mn	76,900	99,300	94,400	89,100	50,000	30,600	27,200	36,600	35,700	35,600	32,400
4.12% Mn	84,900	107,200	109,000	103,000	79,000	34,200	43,200	45,000	51,000	35,800	39,700
5.10% Mn	93,000	113,300	117,200	107,500	91,600	33,800	53,900	49,700	49,100	57,700	64,900
5.85% Mn	102,700	115,800	119,800	106,300	91,000	37,400	32,600	67,700	44,000	51,400	58,400
6.55% Mn	122,000	125,000	124,700	113,200	91,600	44,600	.....	63,600	57,500	60,000	63,200
6.95% Mn	115,200	131,600	126,200	116,500	95,800	46,400	38,900	73,400	60,600	60,300	66,100
		Specimens heated 8 hours at indicated temperature, air-cooled, and reheated 1 hour at 1150° F.									
Armco	89,200	.....	.....	.....	.....	.....	.....	.....	.....	.....	.....
1.15% Mn	94,100	.....	.....	.....	.....	36,600	32,000	21,400	30,300	31,900	38,200
1.88% Mn	80,400	.....	.....	.....	.....	45,000	30,800	45,300	41,200	42,300	41,300
3.15% Mn	76,900	.....	.....	.....	.....	39,600	31,000	50,300	57,200	53,100	52,800
4.12% Mn	84,900	.....	.....	.....	.....	53,800	65,500	69,300	64,200	62,900	55,100
5.10% Mn	93,000	.....	.....	.....	.....	64,900	77,900	64,500	61,600	63,300	59,600
5.85% Mn	102,700	.....	.....	.....	.....	64,400	83,600	65,400	68,000	66,400	66,300
6.55% Mn	122,000	.....	.....	.....	.....	.....	.....	70,800	69,700	67,500	68,100
6.95% Mn	115,200	.....	.....	.....	.....	62,500	76,800	70,400	72,400	67,600	70,600

Table V
Per Cent Elongation of Cold-Worked and Heat Treated Iron-Manganese Alloys

Alloy	60% C. W.	800	900	1000	1100	1200	1300	1400	1500	1600	1700
		Specimens heated 8 hours at indicated temperature and furnace-cooled — Degrees Fahr.									
Armco	3.3	11.7	....	....	46.0	....	....	....	....	....	....
1.15% Mn	3.1	11.5	13.4	13.8	28.2	42.5	42.7	42.8	41.8	38.3	40.1
1.88% Mn	4.7	6.8	9.5	30.8	35.8	40.0	37.8	36.7	36.4	31.4	37.6
3.15% Mn	4.7	10.5	12.9	20.5	33.5	34.8	22.5	17.4	23.8	24.7	25.5
4.12% Mn	4.4	9.8	12.7	18.3	23.3	24.3	11.5	13.3	13.3	14.8	16.4
5.10% Mn	5.6	9.6	6.8	13.7	23.2	16.5	8.2	9.7	10.6	10.6	10.3
5.85% Mn	3.9	6.2	8.8	17.7	23.5	13.7	7.3	7.7	7.9	8.5	8.2
6.55% Mn	4.3	8.0	12.7	16.8	25.7	11.7	7.2	6.8	7.0	7.5	7.2
6.95% Mn	3.5	6.1	11.8	16.8	23.5	11.3	6.8	6.7	6.6	7.1	6.9
		Specimens heated 8 hours at indicated temperature and air-cooled									
Armco	3.3	11.3	....	....	41.3	....	....	....	....	....	....
1.15% Mn	3.1	9.5	11.9	13.9	31.0	37.0	42.3	39.6	38.9	32.5	33.0
1.88% Mn	4.7	7.7	9.3	12.5	36.0	37.1	37.3	31.2	36.4	30.5	33.5
3.15% Mn	4.7	8.1	11.4	13.8	32.0	34.0	22.2	16.3	14.5	14.8	14.6
4.12% Mn	4.4	8.3	9.3	10.9	25.8	22.0	12.3	10.9	8.2	10.3	9.3
5.10% Mn	5.6	7.3	6.4	9.9	25.2	16.7	8.2	6.5	6.3	8.4	7.2
5.85% Mn	3.9	5.8	7.2	12.5	25.0	14.0	6.3	6.2	6.5	6.0	6.2
6.55% Mn	4.3	7.8	9.3	14.0	24.0	13.3	....	5.6	6.2	5.6	6.0
6.95% Mn	3.5	6.9	11.1	12.5	22.0	11.2	6.0	5.8	6.1	5.9	6.2
		Specimens heated 8 hours at indicated temperature, air-cooled, and reheated 1 hour at 1150° F.									
Armco	3.3	....	....	....	....	....	....	....	....	....	....
1.15% Mn	3.1	....	....	....	....	27.0	40.2	41.0	40.2	35.0	35.7
1.88% Mn	4.7	....	....	....	....	36.2	37.0	36.6	37.1	34.7	34.2
3.15% Mn	4.7	....	....	....	....	32.0	29.8	27.7	22.9	24.2	25.2
4.12% Mn	4.4	....	....	....	....	22.8	24.3	22.5	22.3	21.5	21.8
5.10% Mn	5.6	....	....	....	....	23.4	24.8	21.9	21.4	20.5	21.5
5.85% Mn	3.9	....	....	....	....	23.0	22.5	21.7	21.2	22.5	22.0
6.55% Mn	4.3	....	....	....	....	....	....	19.3	20.8	20.2	19.3
6.95% Mn	3.5	....	....	....	....	22.0	21.2	20.5	19.6	18.9	18.7

Table VI
Rockwell Hardness of Cold-Worked and Heat Treated Iron-Manganese Alloys

Alloy	60% C. W.	800	900	1000	1100	1200	1300	1400	1500	1600	1700
		Specimens heated 8 hours at indicated temperature and furnace-cooled — Degrees Fahr.									
Armco	96B	88B	..	..	39B	..	..	..	..	..	..
1.15% Mn	97B	95B	91B	83B	52B	50B	46B	47B	50B	47B	47B
1.88% Mn	99B	99B	94B	77B	62B	60B	53B	58B	63B	54B	53B
3.15% Mn	101B	102B	97B	81B	72B	73B	89B	95B	84B	77B	75B
4.12% Mn	26C	26C	22C	15C	12C	6C	22C	17C	16C	14C	14C
5.10% Mn	28C	29C	25C	19C	19C	17C	27C	23C	22C	19C	22C
5.85% Mn	32C	32C	26C	22C	18C	23C	29C	25C	23C	22C	21C
6.65% Mn	34C	34C	28C	26C	23C	28C	28C	29C	26C	26C	29C
6.95% Mn	37C	35C	30C	27C	23C	29C	32C	29C	28C	27C	29C
		Specimens heated 8 hours at indicated temperature and air-cooled									
Armco	96B	90B	..	..	49B	..	..	..	..	..	..
1.15% Mn	97B	98B	95B	90B	54B	57B	43B	48B	55B	64B	57B
1.88% Mn	99B	99B	100B	94B	63B	64B	53B	67B	63B	65B	63B
3.15% Mn	101B	99B	100B	96B	70B	74B	86B	93B	90B	89B	85B
4.12% Mn	26C	27C	24C	19C	9C	9C	20C	20C	19C	18C	15C
5.10% Mn	28C	29C	27C	24C	15C	19C	24C	24C	24C	20C	23C
5.85% Mn	32C	32C	29C	26C	18C	21C	26C	26C	27C	23C	27C
6.55% Mn	34C	32C	31C	27C	24C	26C	..	30C	29C	27C	28C
6.95% Mn	37C	34C	31C	29C	24C	28C	29C	29C	30C	27C	29C
		Specimens heated 8 hours at indicated temperature, air-cooled, and reheated 1 hour at 1150° F.									
Armco	96B	..	..	..	..	..	..	..	..	..	..
1.15% Mn	97B	..	..	..	..	52B	45B	40B	54B	55B	56B
1.88% Mn	99B	..	..	..	..	61B	56B	63B	67B	64B	61B
3.15% Mn	101B	..	..	..	..	73B	77B	78B	81B	82B	79B
4.12% Mn	26C	..	..	..	..	87B	91B	88B	88B	87B	84B
5.10% Mn	28C	..	..	..	..	94B	95B	92B	91B	89B	86B
5.85% Mn	32C	..	..	..	..	96B	98B	94B	94B	92B	92B
6.55% Mn	34C	..	..	..	..	..	..	18C	16C	14C	17C
6.95% Mn	37C	..	..	..	..	20C	19C	20C	18C	18C	17C

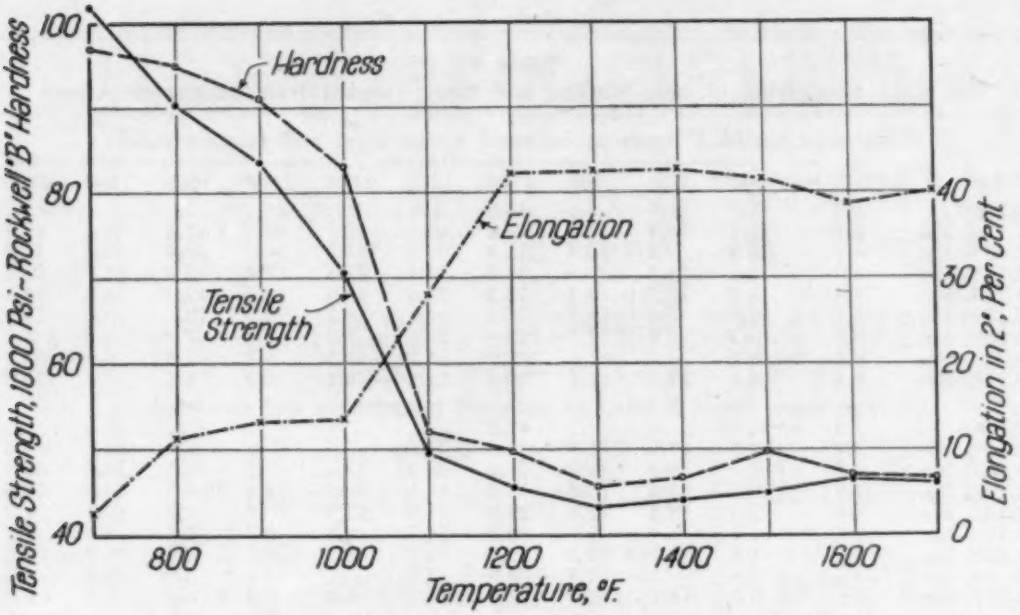


Fig. 13—Physical Properties of a 1.15 Per Cent Manganese Alloy in the Form of $\frac{1}{8}$-Inch Sheet. The specimens were given a 60 per cent reduction in thickness by cold rolling and then annealed at the temperatures indicated.

that the alloys fall into three groups according to the trends of the physical properties.

The first group is characterized by a decrease in strength and hardness and an increase in elongation with increasing annealing temperature. These are the normal effects in single-phase metals and are well-represented by the 1 per cent manganese alloy whose properties are plotted in Fig. 13 as a function of the annealing temperature. Here the strength and hardness fall rapidly for annealing temperatures up to 1100 degrees Fahr. (595 degrees Cent.) and then change very little as a result of higher annealing temperatures. The elongation behaves in an analogous manner, reaching a high point after a 1200 degree Fahr. (650 degree Cent.) anneal, and remains substantially constant with higher temperatures. These property changes are in complete accord with the normal strain relief, recrystallization, and grain growth usually expected as a result of annealing. The 1 and 2 per cent alloys behave in this manner. According to the diagram of Troiano and McGuire (5), these two alloys fall into the single-phase alpha solid-solution field for all temperatures up to 1400 degrees Fahr. (760 degrees Cent.), and at 1600 and 1700 degrees Fahr. (870 and 925 degrees Cent.) they become completely austenitic. The martensitic constituent formed from gamma on air cooling from 1500 to 1700 degrees Fahr. (815 to 925

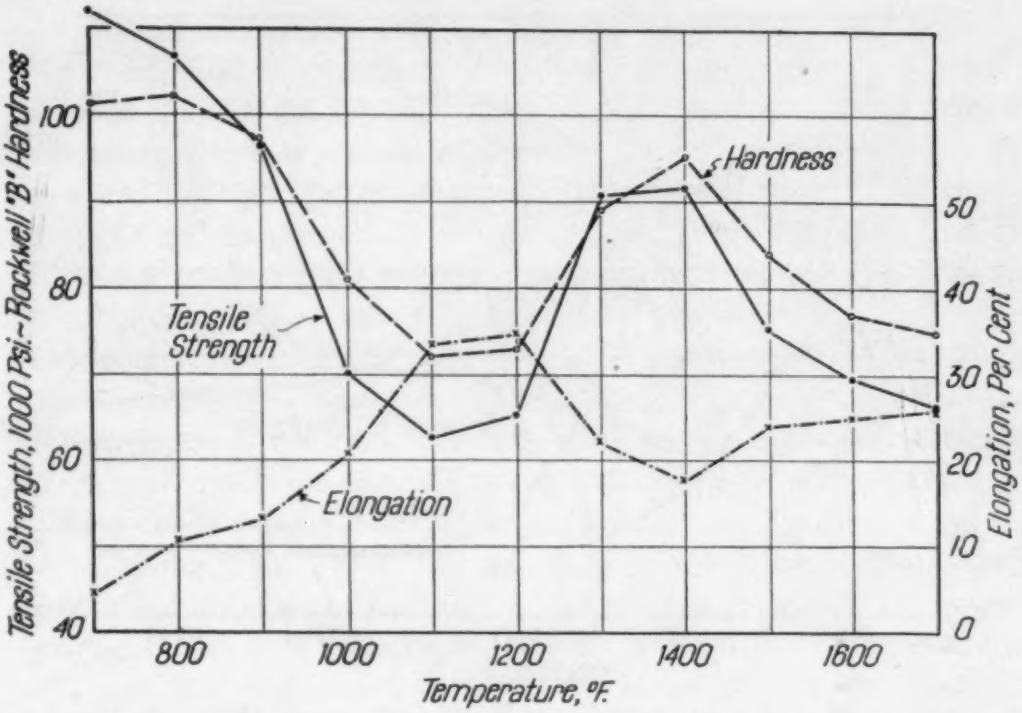


Fig. 14—Physical Properties of a 3.15 Per Cent Manganese Alloy in the Form of ⅛-Inch Sheet. The specimens were given a 60 per cent reduction in thickness by cold rolling and then annealed at the temperatures indicated.

degrees Cent.) is apparently too small in amount or too low in manganese content to have much effect on the properties, and the alloys behave in a normal manner.

The second group is characterized by a secondary hardening that occurs in material annealed in the temperature range of 1200 to 1400 degrees Fahr. (650 to 760 degrees Cent.). These annealing temperatures produce an abnormal increase in hardness and tensile strength, with an accompanying decrease in elongation. The changes reach their peaks at 1300 to 1400 degrees Fahr. (705 to 760 degrees Cent.) and are reversed with higher annealing temperatures, the properties then approaching those obtained by 1100 to 1200 degree Fahr. (595 to 650 degree Cent.) annealing. The group is represented by the data for the 3 per cent alloy plotted in Fig. 14. In this graph the strength and hardness decrease, showing a minimum, while the elongation reaches a maximum with an annealing temperature of 1100 to 1200 degrees Fahr. (595 to 650 degrees Cent.). At higher annealing temperatures the first two properties increase to a maximum and the elongation drops to a minimum with annealing at 1400 degrees Fahr. (760 degrees Cent.). Still higher annealing temperatures cause the strength and hardness to decrease and the elon-

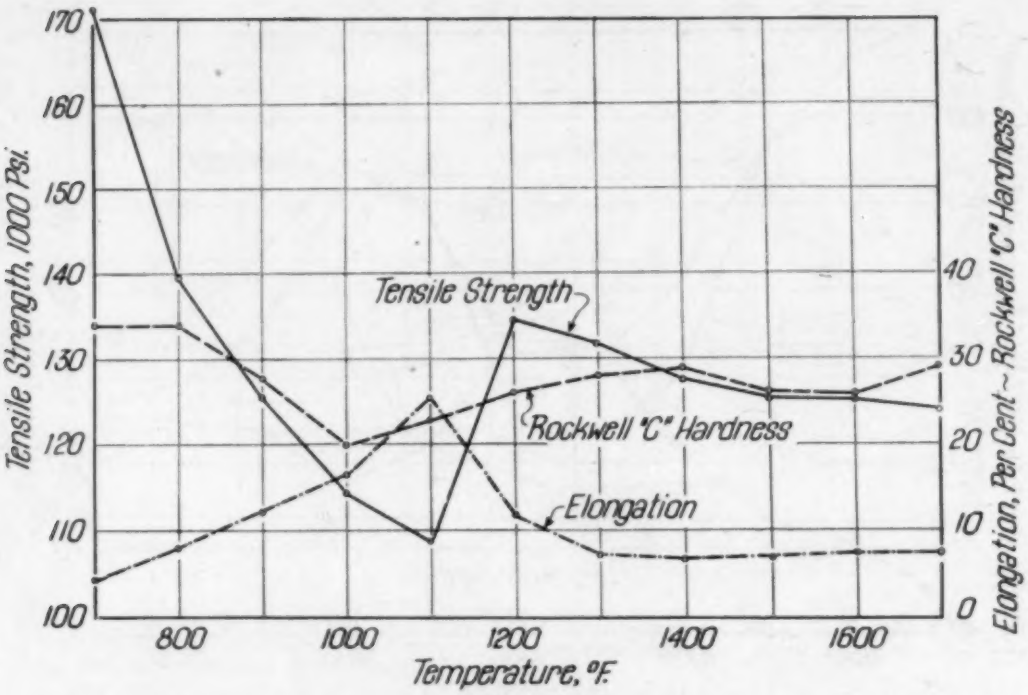


Fig. 15—Physical Properties of a 6.55 Per Cent Manganese Alloy in the Form of $\frac{1}{16}$-Inch Sheet. The specimens were given a 60 per cent reduction in thickness by cold rolling and then annealed at the temperatures indicated.

gation to increase, with the individual properties approaching the values attained with the 1100 degree Fahr. (595 degree Cent.) anneal. The abnormal property changes produced by annealing in the range of 1200 to 1400 degrees Fahr. (650 to 760 degrees Cent.) is no doubt due to a secondary hardening resulting from formation of the gamma solid solution with a relatively high manganese content. The decomposition of this phase into a martensitic constituent (Troiano's supersaturated alpha) of high strength and hardness readily accounts for the observed properties. Still higher temperatures promote the formation of greater amounts of the gamma solution with decreasing manganese content and the martensitic constituent formed during the cooling would be expected to have lower strength and hardness thus producing the observed decrease in properties. Alloys containing 3 to 6 per cent manganese fall into this group and the behavior postulated above is quite consistent with the accepted alpha to gamma changes for these compositions.

The third group differs from the second in that the hardening effect persists after annealing at all temperatures beyond 1200 degrees Fahr. (650 degrees Cent.). The hardening sets in after annealing at 1200 degrees Fahr. (650 degrees Cent.), and the strength

and hardness increase with an appropriate decrease in elongation. With still higher annealing temperatures only minor decreases in strength and hardness occur. The 6 and 7 per cent alloys fall into this group which is well represented by Fig. 15 where the properties observed for the 6 per cent alloy are shown. The initial decrease in strength and hardness and increase in elongation follow the pattern of group 1 while the secondary hardening is consistent with group 2. The properties remain substantially constant with still higher annealing temperature because the alloys cross over into the single-phase gamma field of the diagram, and since the manganese content of this phase is constant with increasing temperature the properties of its martensitic decomposition product would be quite constant for similar conditions.

Cooling the alloys in still air, as previously mentioned, produces properties only slightly different from those obtained by furnace cooling. This is in accordance with the results of other investigators (6) and was also noted during the experiments to determine the proper annealing temperature for cold-worked material. While the alloys were divided into three separate groups for discussion of their annealed properties, it is obvious that they are closely related and that one group merges into the other with no sharp division between them. The properties of any given alloy in this range of composition are necessarily a function of its manganese content, its mechanical history, and its thermal history. They deviate from the expected trends only because of the unusual behavior of the face-centered cubic iron-manganese solid solution. The instability of this phase on cooling and the low temperatures at which it decomposes give rise to the anomalies observed.

Alloys tempered at 1150 degrees Fahr. (620 degrees Cent.) after air cooling from temperatures in the gamma solid solution range show significant decreases in tensile strength and hardness with simultaneous increases in elongation. This is particularly pronounced in alloys containing more than 3 per cent manganese. It would appear that the alloys prior to tempering are essentially martensitic in character as a result of instability of the gamma solution. The tempering alters this martensitic condition thereby decreasing the hardness and strength and increasing the elongation. The exact nature of the changes cannot be followed readily by metallographic means but probably involves relief of strain induced by the transformation and the formation of greater amounts of the equilibrium alpha-solution

phase. This is substantiated by the relative insensitivity of the properties to the normalizing temperature, when this treatment is followed by tempering.

Tempering at 1150 degrees Fahr. (620 degrees Cent.) gives practically identical properties despite wide variation in the normalizing temperature. The hardness and tensile strength are slightly less and the elongation is practically the same as those produced by holding 8 hours at 1100 degrees Fahr. (595 degrees Cent.) and furnace or air cooling. Apparently all three treatments closely approach the same stable conditions.

Summary

Iron-manganese alloys containing up to 7 per cent manganese have been shown to be amendable to cold working. They have been cold-rolled to 80 per cent reduction in thickness and the properties resulting from this cold deformation determined on $\frac{1}{16}$-inch sheet. The 1 and 2 per cent alloys could be readily rolled with relatively heavy reductions, while the 6 to 7 per cent alloys, because of their greater strength and hardness, were reduced by much lighter reductions per pass. Tensile properties were determined in various stages of cold reduction and heat treatment after the cold work.

The properties in the normalized and tempered state indicate a regular increase in tensile strength of about 11,000 pounds per square inch for each 1 per cent manganese with an initial decrease of 5 per cent in elongation tapering down to about 1 per cent for each per cent of manganese at the higher manganese contents while the hardness varies from Rockwell B-55 to C-23. The 7 per cent manganese alloy has a tensile strength of 114,900, a yield strength of 70,000, and a proportional limit of 56,000 pounds per square inch, with an elongation of 16 per cent in 2 inches and a hardness of Rockwell C-23.

In the cold-worked condition the tensile strength is doubled by 80 per cent reduction of the low-manganese alloys and increased by about 80 per cent in the high-manganese alloys. The 7 per cent alloy shows a tensile of 192,500 pounds per square inch in this state. The elongation is greatly affected by the initial reductions but decreases more slowly with additional working and is changed but little over the range of 40 to 80 per cent reduction. Manganese increases the elongation of heavily cold-worked material by a few per cent for

initial manganese additions after which the elongation remains constant. That is, the 3 to 7 per cent alloys have the same elongation at each cold working level.

Annealing and normalizing treatments at temperatures up to 1100 or 1200 degrees Fahr. (595 or 650 degrees Cent.) produce the usual softening of cold-worked material with greatest effects in this temperature range. In alloys from 3 to 7 per cent manganese higher temperatures will produce a hardening effect, due to the formation of a martensitic constituent resulting from the decomposition of the iron-manganese gamma solid solution. The properties of the alloys so treated will vary with the manganese content and the heat treating temperature.

A tempering treatment of 1150 degrees Fahr. (620 degrees Cent.) superimposed on previously normalized material tends to produce substantially constant properties independent of the normalizing temperature and approximating those obtained by normalizing from 1100 degrees Fahr. (595 degrees Cent.). Apparently the 1150 degree Fahr. (620 degree Cent.) tempering temperature is capable of producing the softest condition in all of the alloys by modification of the martensitic constituent formed by the other treatments and by the production of a maximum amount of the ferrite solution.

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DISCUSSION

G. F. COMSTOCK:[2] I was very much interested in the statement that 2 per cent manganese in these iron-manganese alloys decreased the yield strength. I would be interested in the authors' comment on whether this might be due to isolation of nitrogen, oxygen or sulphur or what the reason is according to his best information.

R. E. CRAMER:[3] We were very much interested in the manganese alloys but our work has to do with railroad rails, and we would be interested in knowing the effect of carbon on these manganese alloys. We hope the authors in some of their future work will go on and add carbon to medium manganese and give us a wide series of results.

Authors' Reply

We appreciate Mr. Comstock's interest in the lower yield point with the 2 per cent alloys and at the same time the higher tensile strength which will be indicative of an alloy adaptable to working processes. At this particular moment we are not capable of answering the question as to whether oxides or some other impurity function is contributing to the decrease at that point. These alloys were made by direct additions without any efforts to control those factors and they have not been completely determined. In seven successive heats, in examining that particular composition we found similar results and we are endeavoring to isolate the reason. At the moment we do not know.

The carbon content in manganese alloys, as Professor Cramer requests, has to some extent in the heat treated condition been covered by several other authors. However, they have not maintained the lowest silicon and phosphorus values and we hope at some future date to give you that particular information.

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[3]Special research associate professor of engineering materials, University of Illinois, Urbana, Ill.

STRESS COMPARISONS BY CORRELATION WITH HIGH FREQUENCY MAGNETIC AND EDDY CURRENT LOSSES

By P. E. Cavanagh

Abstract

This paper examines the possibility of using a high frequency oscillator, whose output is governed by magnetic and eddy current losses, to accomplish practical comparisons of stresses in metals.

The success obtained by various investigators in recent years in comparing and measuring internal stresses by magnetic means is mentioned.

Some preliminary experiments are described to establish the fact that high frequency core losses do correlate with internal stresses.

Practical applications are divided into (a) comparisons of stresses arising from cold working and quenching and (b) prediction of fatigue failure. The advantages of this method are obvious. The practical difficulties encountered are emphasized.

Introduction

THE material contained in this paper is mainly a recital of preliminary work in a number of new fields of investigation. The results have been encouraging enough to warrant further work which is now in progress. The main purpose of this present account is to define clearly the limitations of this method of stress comparison and to indicate those applications having the most practical value in the light of our present knowledge.

The theoretical and experimental establishing of useful correlations between magnetic properties and internal stresses has demonstrated the possibilities of developing a nondestructive method of determining stress conditions in metals. The agreement recently reached between fundamental magnetic theory and experimental re-

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sults has allowed the theoretical explanation of many phenomena which formerly raised doubts as to the validity of using magnetic determinations in this manner (1).[1]

Any force which causes a distortion of the crystal lattice will produce some effect on the magnetic properties of a ferromagnetic material. Whether this change is measurable or not is a practical problem. The theoretical relationships between various types of stresses and magnetic properties have been established by several investigators (2), (3), (4), (5). Usual methods of determining magnetic properties accurately are not rapid and are usually destructive if used on production material. A description of the successful development of a production tool for measuring internal stresses in ferromagnetic material and a specific application of the method to nickel wire is contained in a series of papers by F. Förster (6), (7). By measuring the change in remanence caused by small increments of tensile stress, the internal stresses can be calculated from the relationship—

$$\left(\frac{d J_r}{d\sigma}\right)_{\sigma \to o} = \frac{J\infty}{4\sigma_i}$$

where J_r = Intensity of Magnetization at Remanence
$J\infty$ = Intensity of Magnetization at Saturation
σ_i = Internal Stress
$d\sigma$ = An increment of tensile stress small compared to σ_i (2)

As stress is increased in a ferromagnetic material, hysteresis losses and resistivity increase. These are well known and established facts as are the accompanying decreases in permeability, remanence and coercive force for a ferromagnetic material with positive magnetostriction. If stresses are applied to a material which is not ferromagnetic, the electrical resistivity is the only one of these particular stress variable properties left to change.

High Frequency Oscillator

If a metal sample is placed in the field of a test coil which is part of the tuned circuit of an oscillator, energy losses will occur in the metal sample which will decrease the energy output of the oscillator. Both magnetic and eddy current losses will make up the total "core losses" in the coil. Thus, if hysteresis losses, for example, increase in the sample, the oscillator energy output will decrease.

[1]The figures appearing in parentheses pertain to the references appended to this paper.

This principle offers a possible tool for quick comparisons of stress levels within certain definite limits.

If such an oscillator is to provide useful information, it must be extremely sensitive and yet sufficiently stable for practical purposes. These objectives can best be realized in a frequency range from about 1000 cycles per second to about 200,000 cycles per second. In this range, the flux set up in a metal sample will not be uniform across the cross section but will exhibit considerable "skin effect". Using the very low field strengths found in a low power oscillator coil and an extremely sensitive oscillator, the "effective depth of flux penetration" at which a significant change in core losses can be detected will be greater than the depth usually considered as the limit of flux penetrations. This effective depth of flux penetration is governed by the test flux frequency, metal sample permeability, resistivity and diameter.

The distance from the surface of any sample which defines the volume of material in which 86.5 per cent of the core losses occur is approximately expressed by

$$P = \frac{1}{2\pi} \times \frac{\sqrt{\rho 10^9}}{\sqrt{f\mu.2}} = \frac{1405\sqrt{\rho}}{\sqrt{\mu}} \times \frac{1}{\sqrt{f}} \text{ inches}$$

where ρ = specific resistance in ohms/cm^3
μ = permeability at low field strength
f = test frequency

The percentage of the total sample volume which is taken into consideration will be proportional to P and to the radius of the sample.

$$\text{Effective Volume } V = K \times \frac{P}{r}.$$

K is a constant depending partly upon the sensitivity of the detecting instrument to changes in energy level. The more sensitive the instrument, the greater the effective depth of flux penetration at which differences in magnetic and electrical properties can be detected. If flux distribution is identical to a certain depth, small changes in distribution beyond that depth can be discerned if the instrument used is sensitive enough to indicate the resulting differences in energy losses. "Skin effect" allows the examination of the properties of the metal sample to different depths depending on the test frequency chosen.

As stresses are increased in the elastic range, hysteresis losses and resistivity both increase (Fig. 3). Increasing hysteresis losses

increases the core losses. The increase is rapid until complete alignment of magnet domains is attained and slower from that point on.

Increasing resistivity decreases core losses. The EMF causing eddy currents remains constant so that—

$$\text{Eddy current losses } W_E \propto \frac{E^2}{R} = \frac{\text{Constant}}{R} \text{ for a particular test.}$$

The net result of these two opposing tendencies will be an increase or decrease in core losses depending on the relative rates of change of these two stress variable properties with increasing stress. Since these rates of change are different for different materials, and for different initial conditions in the same material, widely different curves are obtained for different analyses and treatment histories of steels when core losses are plotted against stress (Figs. 4, 6 and 7). The higher the test frequency used, the greater will be the relative effect of eddy current losses on total core losses.

In nonmagnetic materials, increasing stresses will result in decreasing core losses, since there are no hysteresis losses to counteract this tendency.

For a certain increase in stresses, the effect of the change in flux penetration caused by the increase in resistivity (ρ) and the decrease in permeability (μ) is usually of minor significance. The result of increasing flux penetration is to include more material in the effective field and increase core losses. This core loss increase will be proportional to the square roots of resistivity and permeability.

$$W_{Total} \propto \frac{\sqrt{\rho}}{\sqrt{\mu}} \text{ if changes in core losses per unit volume are neglected.}$$

If there is a complete balance between changes in core losses per unit volume with increasing stress due to the factors noted before, the accompanying change in flux penetration will cause an increase in total core losses with increasing stress.

The Instrument

The cyclograph is an instrument developed for rapid comparisons of core losses. It fulfills the requirements of sensitivity, stability and flexibility of choice of test frequencies for a practical production inspection instrument. The cyclograph was used to obtain all the data reported.

Fundamentally, the cyclograph consists of a carefully constructed 2-stage oscillator (Fig. 1). The test coil which is part of the tuned

circuit is bifilar wound to obtain maximum mutual inductance between the two sections of the oscillator and to provide self-shielding from outside interference. This coil controls the operating frequency of the oscillator.

The oscillator is essentially a multivibrator. A sensitivity control governs the amount of regenerative feedback and allows adjustment of the initial energy output of the oscillator to any desired level.

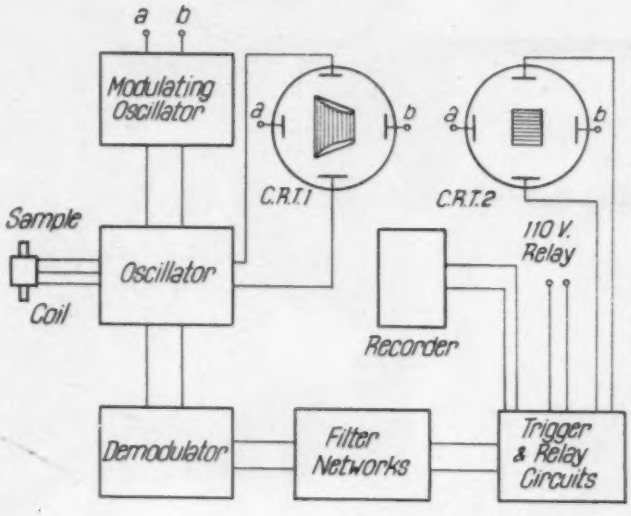

Fig. 1—Cyclograph Block Diagram.

Core losses occurring in the field in the test coil decrease the output of the oscillator. The operating frequency remains constant within narrow limits. In order to stabilize the oscillator when it is operating near zero output, it is modulated by a much lower frequency.

The oscillator output is viewed on a cathode-ray tube screen (Fig. 2). Descriptions of the cyclograph and its uses may be found in current technical literature (8), (9).

Experimental Correlation of Stress and Core Losses

An Esterline-Angus recording milliammeter with 1-mil movement was used to record cyclograph output during various stress cycles.

Curves obtained for steel and aluminum in tension, torsion and compression are shown in Fig. 3. The samples were placed in tensile testing machines in the normal manner with a cyclograph coil around them during the test.

The effects of varying the test frequency are demonstrated in

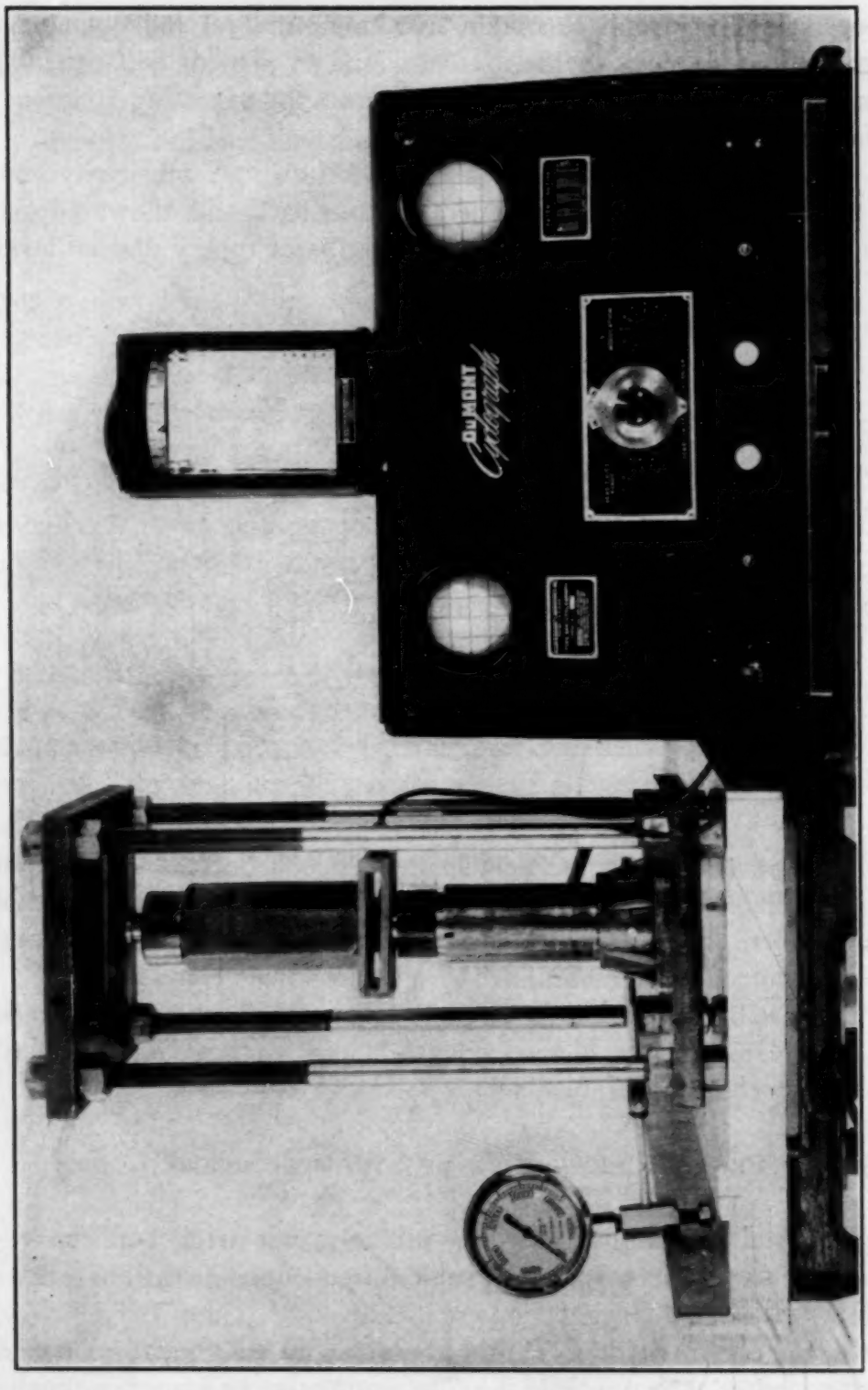

Fig. 2—Cyclograph with Recorder and Small Press for Testing Cartridge Cases.

Figs. 4 and 5 showing curves for steels at different frequencies.

Some representative curves showing that the cyclograph can trace the changes in behavior of magnetic properties with various stress cycles are shown in Fig. 6. The difference in behavior of magnetically "hard" and "soft" materials is clearly illustrated.

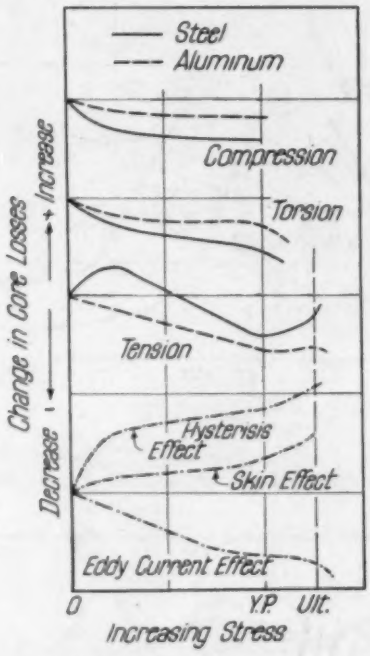

Fig. 3—Effect of Stresses on Total Core Loss.

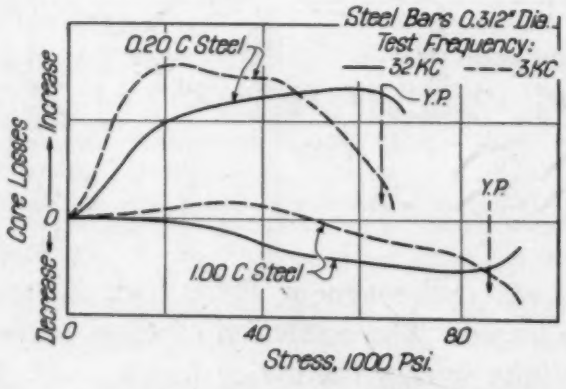

Fig. 4—Correlation Between Stress and Core Losses at Two Frequencies.

The rapid increase in hysteresis losses as the magnetic domains are brought into alignment by external stress causes an initial increase in core losses if the domains are initially in a state of random orientation. When alignment is completed by stresses in the elastic

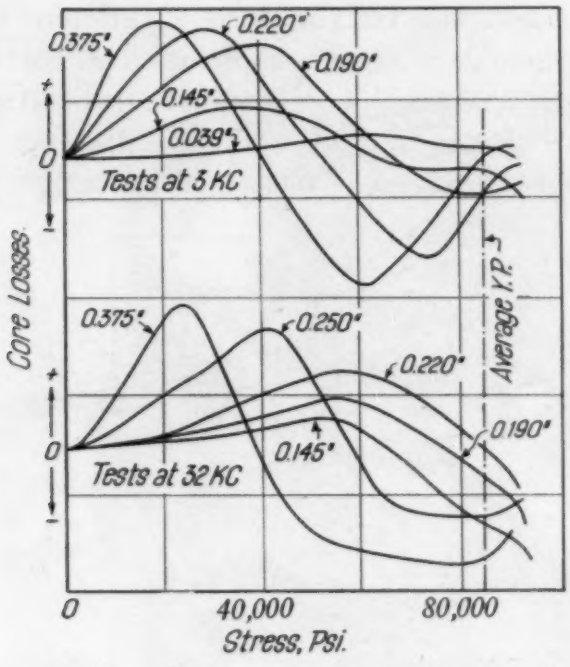

Fig. 5—One Per Cent Carbon Drill Steel Rods, Various Diameters, Two Frequencies.

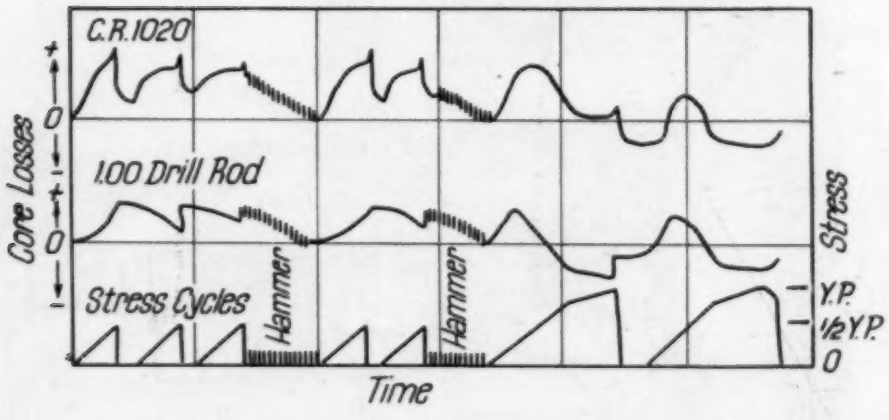

Fig. 6—Difference in Behavior of Core Losses in Two Steels During the Same Stress Cycles.

range or by an external magnetic field, further stressing results in decreasing core losses. The reduction of eddy current losses due to increasing resistivity is now the major factor.

The difference between the partial magnetic recovery of a low carbon steel, and the change in magnetic properties of the high carbon steel when stresses in the elastic range are removed is clearly shown. Hammering the samples after stressing brought them back to a condition of random orientation of the magnetic domains.

If the samples are stressed beyond the yield point, the local

internal stresses set up are not oriented. These random stresses destroy the alignment of the magnetic domains as effectively as hammering does in the elastic range. This is shown by the last stress cycle in Fig. 6.

The curves in Fig. 7 were taken for cold-rolled S.A.E. 1020 as-rolled and the same steel after heating to 600 degrees Cent. (1110 degrees Fahr.) and cooling slowly.

The behavior of the stress-relieved sample shows the expected variations in core losses with increasing stress. In a cold-rolled rod,

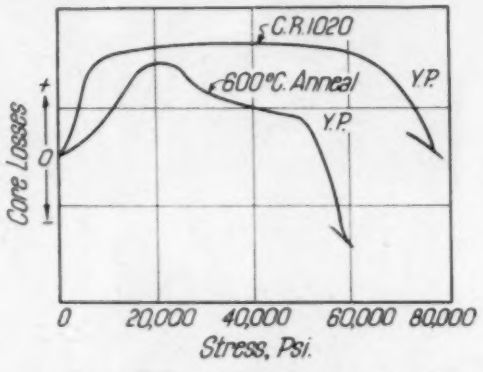

Fig. 7—Core Loss Curves of S.A.E. 1020 Steel As-Rolled and After Heating to 600 Degrees Cent. with Slow Cooling.

the surface layers of the metal are under tension and a central core is under corresponding compression. As a tensile stress is applied, the amount of material under tension increases until finally all the material is under tension.

As stresses increase, flux penetration will increase rapidly, taking into account more and more material which is passing from a neutral zone of zero stress to an initial small tensile stress. The hysteresis losses, and therefore the core losses in this transition zone will increase very rapidly with small increments of tensile stress. This effect counterbalances the normal decrease in core losses appearing in the unstressed sample above 20,000 pounds per square inch.

The response characteristics of the cyclograph circuits must be kept in mind. Over the range of frequencies used, there is no significant difference in response. It is not always possible to set the instrument to the same sensitivity for different materials and sample sizes so that results are generally comparable only over a limited range. For instance the significance of a certain change in output in terms of stress on one of these figures is not necessarily the same as in the others. Since all results are comparative, any

Fig. 8—Forty-Millimeter Cartridge Case Ready for Test. Case is full of paraffin. Plunger at top applies pressure to paraffin. Cut-away gage confines internal hydrostatic load to base of case. Test coil is placed around zone 1 inch from base of case.

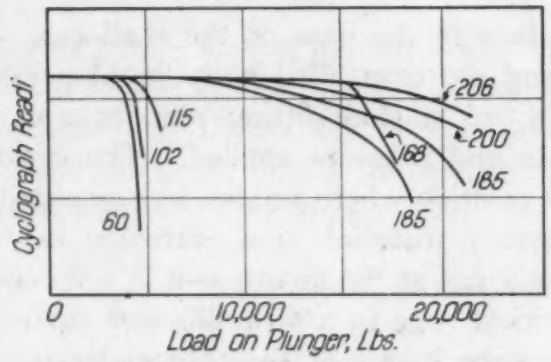

Fig. 9—Forty-Millimeter Cases of Various Hardnesses. Vickers readings are noted.

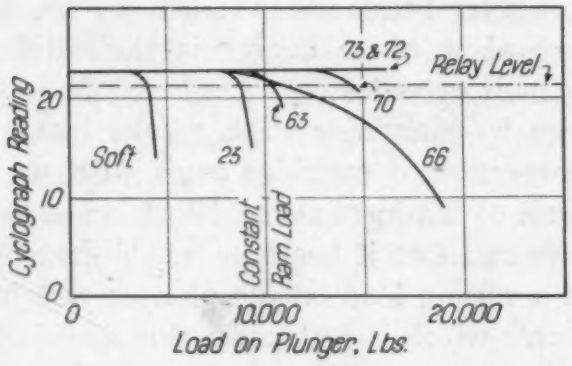

Fig. 10—Forty-Millimeter Cases Loaded to Failure. Rockwell B hardness noted. Possible choice of relay level setting and ram load is indicated which would reject cases below 65 Rb.

setting of the cyclograph may be taken as a reference zero, and the increase or decrease of core losses traced from this reference. Cyclograph readings vary inversely as total core losses.

One possible application of the correlation found between cyclograph output and stress-strain curves is the development of a simple "go-no go" type of tensile or compression test.

Development of a Practical Test

As an example, a test was developed for small caliber brass cartridge cases. In an attempt to devise a test which would approximate conditions during actual firing in a gun, 40-millimeter brass cartridge cases were filled with paraffin and tested under internal pressure.

It has been found in testing other metal shapes that uniform hydrostatic pressure can be obtained by this method.

The primer hole in the base of the shell case was first closed with a T plug and the case filled with liquid paraffin. After the paraffin had solidified, a close fitting plunger was inserted in the mouth of the case and pressure applied. The cyclograph test coil was placed at the position where failure was expected to occur.

Since the softest material in a cartridge case occurs at the mouth, the case will fail at the mouth if it is not confined.

The most critical zone in a 40 millimeter cartridge case occurs about 1 inch from the base. If the case under test is placed in a cut-away gage as shown in Fig. 8, failure can only occur in the critical zone just below the gage.

The results of several tests are included in Figs. 9 and 10. One interesting feature about these curves is their different behavior after the yield point is exceeded.

It can be seen by examining these results that a practical test is available for inspection of cartridge cases. Let us assume that a specified lower limit of hardness at the 1-inch mark has been set for 40-millimeter cases and that it has been established by this type of investigation that a certain load on the plunger will just exceed the yield point for a case which is just under this specified limit.

It is a simple matter to fill a percentage of small caliber ammunition cases with paraffin, place them in a small press as shown in Fig. 2 and apply a specified dead load. Relay circuits in the cyclograph can be set to operate when a certain definite increase in core losses takes place. If the yield point is exceeded, core losses will increase sharply and the cyclograph relays will operate a warning light indicating that the case is below specifications.

If desired, a permanent record of the cyclograph output can be taken on the recorder during each test.

The advantages of this type of test are obvious.

It eliminates the difficulty of cutting representative tensile specimens from small cartridge cases. The test takes the whole circumference into account instead of a local area.

Conditions approximate an actual service test in a gun. Pressure is applied in the same manner but at a much slower rate than in service.

Internal stresses are not relieved in preparing specimens. This is presumably not an important point since circumferential stresses are usually negligible in brass cartridge cases. However, in cyclograph inspection of large brass cases, some cases have been found

where compressive stresses in the wall near the base were of the order of 10 per cent of the yield strength. These would be relieved when tensile specimens were cut and therefore not taken into account.

This type of test might be termed "semi-destructive" since satisfactory cases are not harmed and can be cleaned out and used after testing.

Discussion

The preliminary experiments described establish the fact that total core losses can be used to correlate with stresses in metals. The practical application of stress comparisons to production pieces is quite a different problem. Core losses are affected by a large number of variables such as analysis, heat treating history, grain size, etc. Major differences in core losses are sometimes caused by variations in physical properties which have no significant effect on service life. In order to apply such a correlation to production problems, all extraneous variables must be eliminated. This fact limits the application of the cyclograph in this field to the following instances:

1. Where parts are tested in the cyclograph, then stressed and tested again to find the change in core losses. This procedure can be extended to testing samples during stressing.
2. Where it is definitely known that nothing varies from piece to piece except stresses.
3. Where it is possible to use the cyclograph at one frequency to first separate samples into groups in which the pieces are sufficiently similar in extraneous variables to allow subsequent successful correlation with stresses at another frequency.

Base Caps on 119 Fuzes

One practical application illustrates the necessity of controlling extraneous variables before it can be safely assumed that differences in core losses are due solely to stresses.

In assembling the fuze illustrated in Fig. 11, the base cap must be screwed down tight to crush the tetryl pellet, which is approximately 0.015 inch oversize, in order to obtain proper operation of the fuze. If the pellet is properly crushed, stresses are set up in the metal cap. The cyclograph will distinguish between the stress level in properly loaded fuzes and fuzes where the pellet is either missing or not crushed. In the latter cases, there will be very low stresses in the cap.

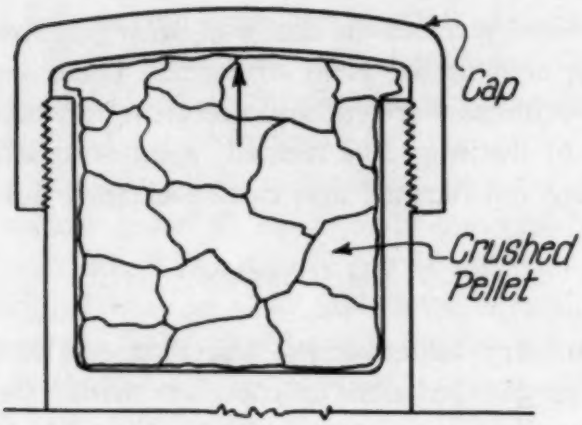

Fig. 11—Stress Developed in Base Cap of Fuze 119.

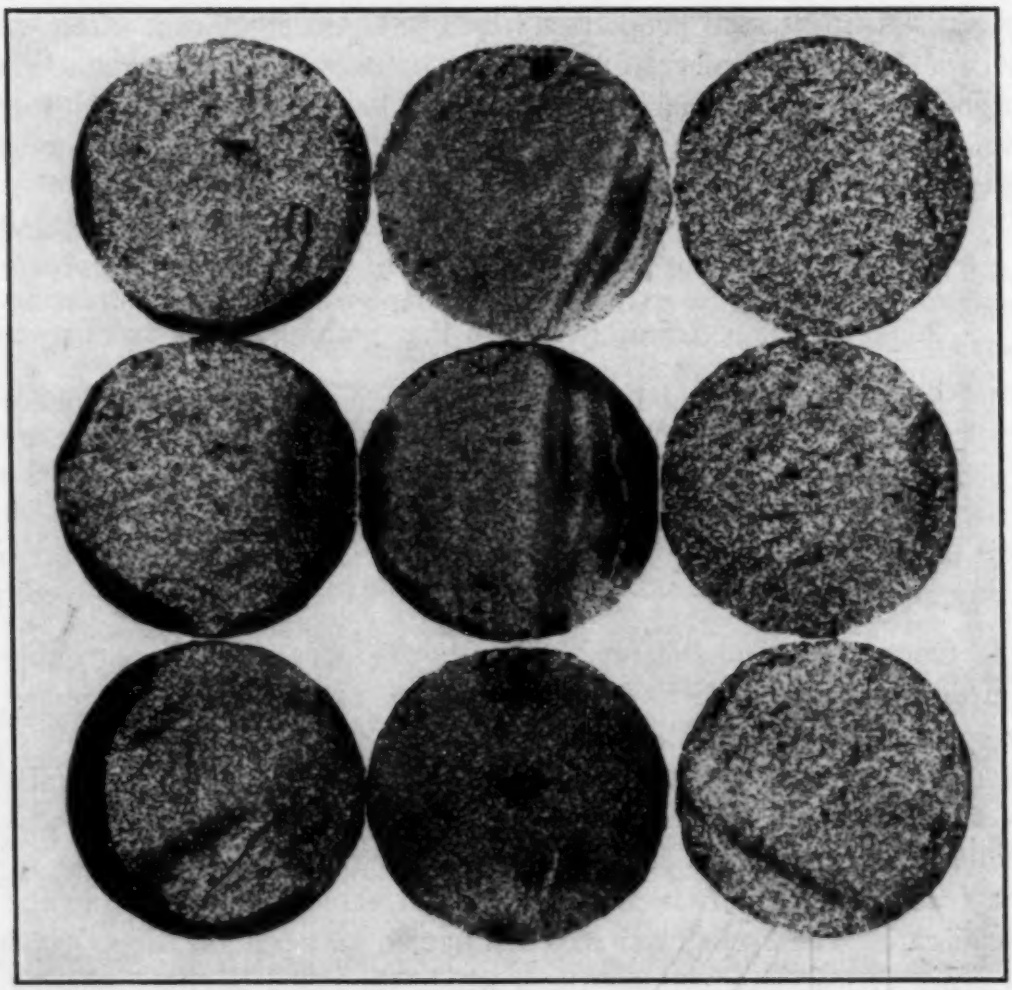

Fig. 12—Fractures of Sten Gun Pins. Left column—circumferential cracks. Low losses. Coarse grain. Center column—fine grain. Satisfactory pins. Right column—Coarse grain. High losses.

In practical application, it was found that wide variations in analysis of the caps made it impossible to rely on this test unless caps of similar analysis were being compared. Keeping caps segregated in lots according to manufacturer proved successful in eliminating this analysis variable.

A 1. Quenching Stresses

Quenching stresses cause cracks. The cyclograph does not indicate the presence of a crack except in certain special instances where a crack may completely break the magnetic and electrical circuit in a closed ring.

High frequency core losses will indicate the presence of high internal stresses which may give rise to cracks. If a piece has cracked badly, some of the stresses have been relieved and the stress level may be the same or lower than for a piece which will give satisfactory service. Fig. 12 illustrates such a case. Group B were satisfactory pieces. Group C had abnormally high core losses. Group A had very low core losses. From examination of the fractures, the only difference between Groups A and C is the serious circumferential quench cracks in Group A which relieved stresses to such an extent that core losses were much lower even than for normal Group B. Both Group A and Group C have a coarser grain than B.

In general, the unpredictable extent to which cracking will relieve stresses makes attempted correlation between core losses and stress very hazardous in such an instance.

The successful practical applications of the cyclograph to problems of this type are few. The instrument has been successful in identifying 57-millimeter monobloc A.P. shot which had such high internal stresses that they would crack spontaneously after being stored some time.

A 2. Cold Working Stresses

Internal stresses due to cold work profoundly affect magnetic and electrical properties (3) (5).

Nothing has been said so far about stresses in nonmagnetic materials. Correlation with resistivity gives excellent results in evaluating stresses in cartridge brass. Fig. 13 shows the variation in core losses for finished 40-millimeter brass cases which were identical in hardness before the final anneal. After annealing at

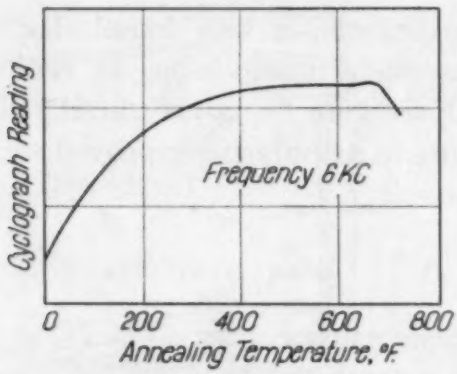

Fig. 13—Forty-Millimeter Brass Shell Cases.

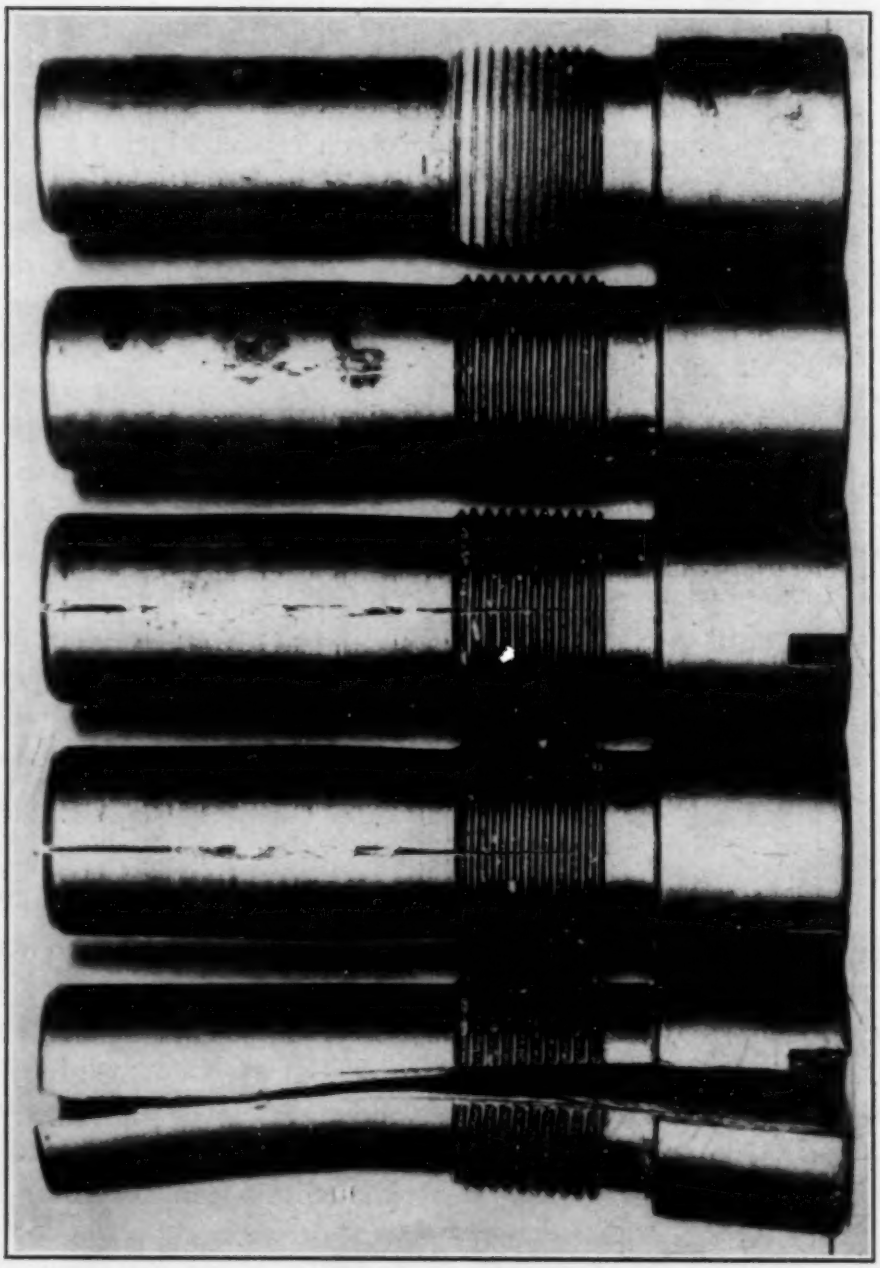

Fig. 14—Various Types of Failure Developed Under Loading Pressure by Tracer 14 Mark II. a—No visible crack before loading. b—Same as a. c—Visible crack before loading. d—Marked bulge on loading. e —Normal satisfactory tracer.

various temperatures, the cyclograph reading changed markedly before any change in hardness or grain size was produced.

In a series of experiments, it was found possible to identify small caliber brass cartridge cases which would crack in the mercurous nitrate test. The surface tensile stresses in cartridge brass must be somewhere between 10,000 and 20,000 pounds per square inch to cause cracking in this test (10). The success of this test to date indicates that brass cases could be segregated by this method into highly stressed cases liable to season cracking and cases with negligible stresses.

40-Millimeter Tracer Bodies

A cyclograph has been in use for some time at the plant of General Engineering Company, Limited, in Toronto, Canada, for the purpose of inspecting 40-millimeter Mark II tracers in production (Fig. 14).

The quantity inspected per day is about 14,000. Results to date have been to reduce the number of splits and accidental fires during loading of the tracers to a very low figure compared to the percentage of splits and fires which occurred before this method of inspection was installed.

The cyclograph is not sensitive to cracks or flaws which may have been present in the original bar stock. The tracers are now being inspected visually for such flaws, but it is not expected that all of this type of defect can be detected by eye. For this reason there will always be some very small number of tracers which may pass both visual and cyclograph inspection and split in loading.

Tracers are loaded with tracer composition under a dead load of 9000 pounds. The stresses developed in the metal approach the yield point for the particular type of high sulphur, free machining steel used. It was first hoped that the cyclograph could be used to reject tracers with low physicals which might be expected to split in loading and that this would eliminate splits. However, it was found that there are several distinct causes for splits and this method did not begin to eliminate all of the defective tracers. Low physicals, segregation of sulphides or slag, local concentrations of high internal stresses, and cracks and flaws in the original bar stock were all found to produce splits on loading. The investigation reported here has aimed at finding some method of cyclograph inspection which will eliminate all of these causes of failure except cracks.

A destructive test was developed to test cyclograph results. The tracers were filled with paraffin and loaded to the breaking point in a press. The pressure required to split a tracer under these conditions indicates whether it will stand the normal loading pressure.

Several interesting facts were brought out by preliminary investigations. If tracers were loaded to a point slightly above the

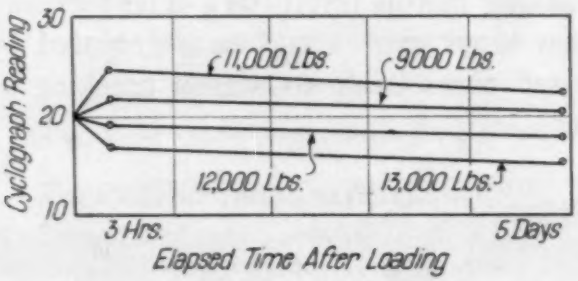

Fig. 15—Recovery in 40-Millimeter Tracers Loaded at Pressures Noted.

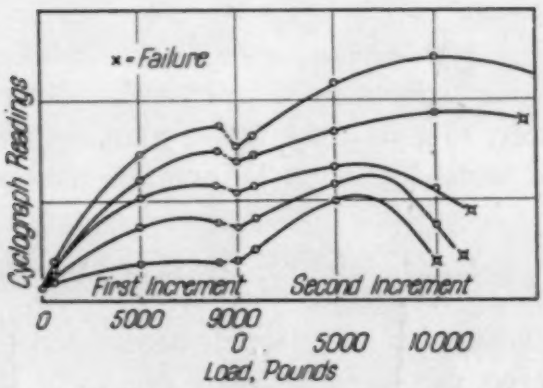

Fig. 16—Bursting Pressure Tests on 40-Millimeter Tracer Bodies.

elastic limit and cyclograph readings taken on them over a period of time, it could be demonstrated that some of the tracers had taken a permanent set while others slowly recovered, relieving some of the internal stresses set up (Fig. 15). If cyclograph readings were taken as a tracer was put under load, satisfactory tracers would exhibit a different relationship between cyclograph reading and applied load than would the unsatisfactory tracers (Fig. 16).

The final method of testing developed was to place the tracers in an eccentric coil and rotate them slowly. A local concentration of high internal stresses would give a very marked variation in cyclograph reading as the tracer rotated.

A crack which was not accompanied by internal stresses gave no indication whatever.

Many of the tracers rejected for high local stresses do not split on loading. A more important fact is that the only tracers passed as having low internal stresses which do split on loading have cracks which were present in the original bar stock.

Stress concentrations are found associated with slag inclusions, sulphide stringers and small cracks apparently developed in cold drawing the bar stock. Other stress concentrations are found which have no apparent correlation with any detectable nonuniformity in structure.

Table I
Barnes-Gibson-Raymond Shot Peened Samples

Shot Peening Time in Minutes	¼-Inch Rods Group No.	Helical Springs 2¼ by 1¼ Inch Diameter Group Number
0	1	1
1	2	.
2	3	2
4	4	3
8	5	4
12	6	5
16	.	6
20	7	7
30	8	8
40	9	.
45	.	9
90	.	10

Shot Peening Stresses

In recent years, shot blasting to increase the fatigue life of springs and other parts undergoing stresses in service has been found to be of considerable value. The peening action of shot on a metal surface sets up beneficial compressive stresses in the surface layers of the metal. The comparison of stress levels in the surface layers of shot peened parts is ideally suited to the high frequency core loss method.

Samples consisted of shot peened rods and shot peened valve springs. The shot peening equipment used was of the mechanical type and all factors such as size of the shot and velocity of shot were held constant. The amount of time the samples were shot peened was intentionally varied to determine the effect on cyclograph readings.

Tests on Shot Peened Rods

For the ¼-inch rods, a frequency of 90,000 cycles was found to give the best correlation with shot peening time. The cyclograph

readings increased with the time of shot blasting (i.e., core losses decreased) (Fig. 17).

Individual rods in each group of samples varied somewhat in cyclograph readings so that no clear separation could be made between adjacent groups in Table I. However, Group 2 could be

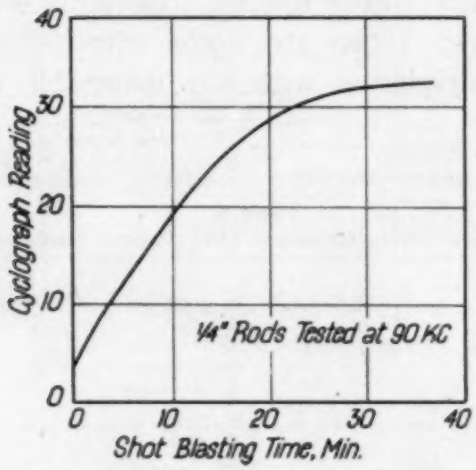

Fig. 17—Cyclograph Curve of Shot Blast Tests of ¼-Inch Rods Tested at 90 KC.

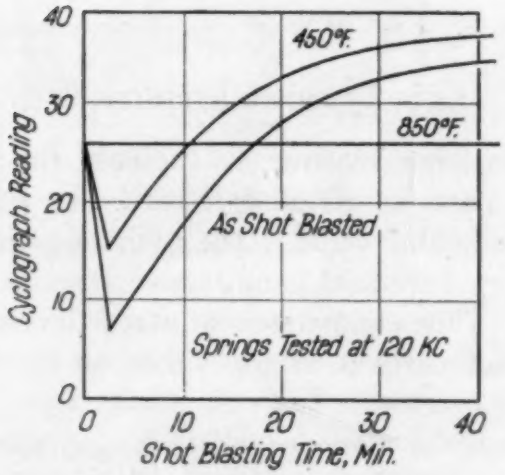

Fig. 18 — Cyclograph Curves of Springs Tested at 120 KC.

separated from Group 4 and Group 3 from Group 5. Similarly every group could be easily distinguished from every other group but those adjacent to it.

The cyclograph readings reached a maximum after 30 minutes' peening. This checks with fatigue tests which show no further improvement in fatigue life after this point.

Test on Shot Peened Springs

Similar groups of springs were tested (Table I) as shot blasted, after stress relieving at 450 degrees Fahr. (230 degrees Cent.) and after heating to 850 degrees Fahr. (455 degrees Cent.). Results of testing springs at 120,000 cycles are shown in Fig. 18.

These results indicate that core losses show definite correlation with shot peening intensity. The only deviation from the results obtained on the rods is that either—

(a) Springs with no shot peening read much higher than expected because of coiling stresses. This premise was disproved by comparing "as-coiled", and completely stress relieved springs, with no shot peening. The difference in core losses due to coiling stresses did not begin to account for the displacement of the zero shot blast samples on the curve.

or (b) Samples with small amounts of shot peening are erratic with respect to the curves obtained for the rods. This is logical since the stress distribution pattern present in the extreme outer layers of the coiled springs with small amounts of shot peening cannot be duplicated in the rods.

The 120-kilocycle test field used penetrates an extremely small distance into the metal samples. At some degrees of shot peening intensity, a balance of tension coiling stresses and compressive peening stresses will be reached in this zone which will give the same cyclograph reading as obtained for a spring with no shot peening. This is the probable explanation for the curve dropping to a minimum at 2 minutes or less shot blasting time and behaving similarly to the curve obtained for the rods from then on.

The slope of the curve decreases with shot blast time which is reasonable since a maximum stress condition should be reached at which little or no effect is produced by further shot peening.

Other attempts to correlate with shot peening intensity met with varied success.

On shot peened articulating rods, good correlation was obtained with Almen numbers for the rods.

On shot peened valve spring washers, no correlation was obtained between core losses and shot blast pressure. In this series of tests, time of peening was kept constant and pressure varied in a Wheelabrator machine.

B. Prediction of Fatigue Failure

Magnetic and electrical properties are extremely sensitive to the beginning of plastic deformation. Förster has detected irreversible deformation far below the yield point in nickel wire by measuring magnetic properties. His work indicates slip occurring in the (111) plane in nickel wire at very low tensile load (7).

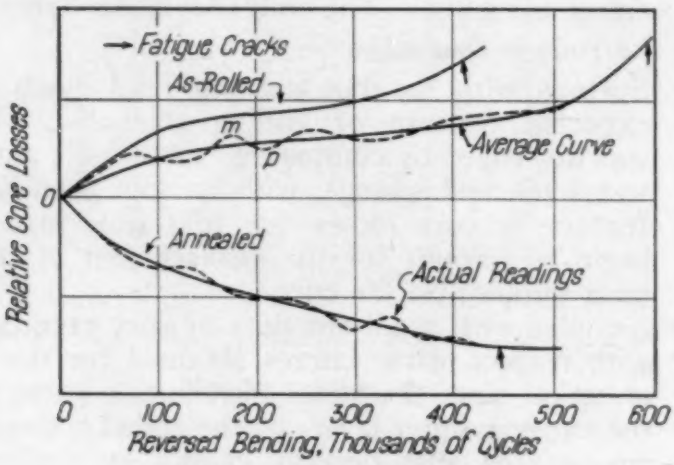

Fig. 19—RPI Fatigue Tests.

An investigation by T. Wlodek (11) traces the changes in the characteristics of the induced voltage generated in a test coil by the magnetization cycle of a specimen subjected to unidirectional repeated stresses. The beginning of plastic deformation is easily detectable by this method.

Table II
R.P.I. Fatigue Test Specimens

Dimensions 1⅛ by ⅛ by 14⅜ Inches

	As-Rolled	Annealed
Modulus of Elasticity	29,500,000 psi.	26,400,000 psi.
Tensile Yield Point	Destroyed by rolling	30,800 psi.
Per Cent Reduction in Area	58.2 Per Cent	44.8 Per Cent
Per Cent Elongation in 2 Inches	8.5 Per Cent	45.0 Per Cent
Ultimate Tensile Strength	106,600 psi.	61,500 psi.
Rockwell Hardness	B96	B68
Carbon Content	0.19 Per Cent	0.19 Per Cent

The cyclograph will also indicate the beginning of plastic deformation. Since fatigue failure usually takes place after plastic deformation has progressed for some time, it is possible to predict fatigue failure in some instances by examining the rate of change of core losses with repeated stresses.

An investigation conducted at Rensselaer Polytechnic Institute

will serve to illustrate some of the possibilities (12). Tests were carried out on specimen strips of steel 1⅛ by ⅛ by 14⅝ inches long both cold-rolled and fully annealed.

Specimens were vibrated magnetically 120 times a second by supporting at the ends and placing a powerful 60-cycle electromagnet below the middle of the strip. Specimens were removed from the fatigue machine and tested after each 50,000 reversals of stress. Stress developed in the specimen was 54,800 pounds per square inch for the cold-rolled material and 28,200 pounds per square inch for annealed material. Table II gives the physical properties of the specimens before testing. The results in Fig. 19 indicate that the behavior of the specimens is a function of their initial condition.

The general trend of the curves can be explained in the same manner as the curves in Fig. 6. The sample is not under external stress when tested. In the annealed sample, increasing resistivity is apparently the controlling factor since core losses decrease with repeated stresses. In the cold-rolled sample, more and more of the original core which was under compression becomes stressed in tension as the test progresses.

In the cold-rolled specimens especially, the core losses oscillate in a regular manner about an average curve. These oscillations smooth out as the fatigue limit of the specimen approaches.

These curves indicate that the effective test field penetrates further into the sample in steps as fatigue progresses. As the point of neutral stress in the specimen moves inward, the field will include a new transition zone where the rate of increase in core losses will be very rapid with further increase in internal stress. After a point such as (m) in Fig. 19, any subsequent change in internal stress in this zone will cause a decrease in core losses. This means that stresses in this particular material are approaching the yield point.

At point (p), a sudden increase in core losses takes place. One factor causing the increase is the penetration of the test field into a new transition zone. A contributing factor may be the appearance of internal stresses above the elastic limit. This cycle of events continues almost until the appearance of microscopic fatigue cracks on the polished surface of the bar. As this point is approached, the oscillations about the average curve decrease in magnitude.

This investigation indicates that the displacement of the stress distribution pattern toward the center of a specimen as fatigue progresses takes place in definite steps, when traced by this method. Each

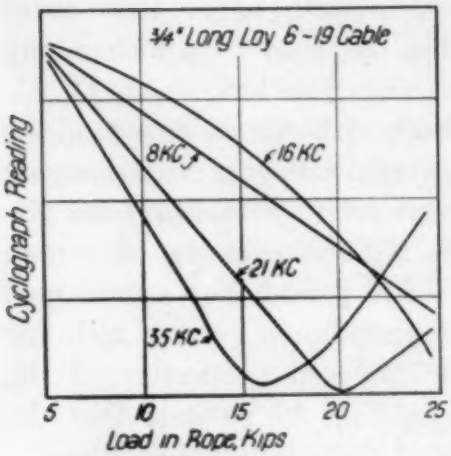

Fig. 20—Effect of Test Frequency on Correlation with Stress in Cable.

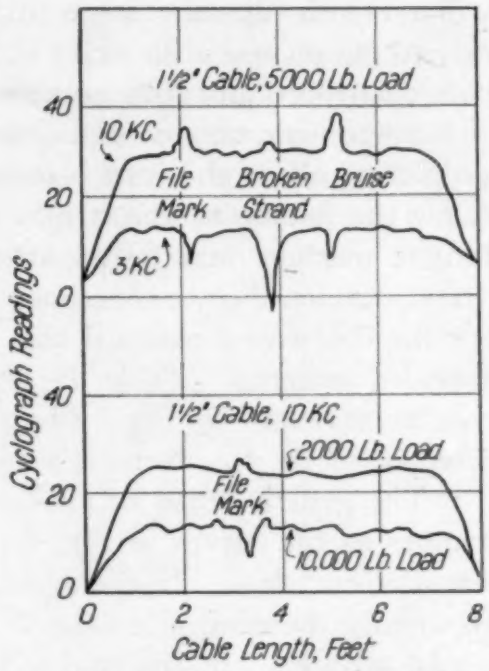

Fig. 21—Effect of Test Frequency and Cable Load on Detection of Cable Defects.

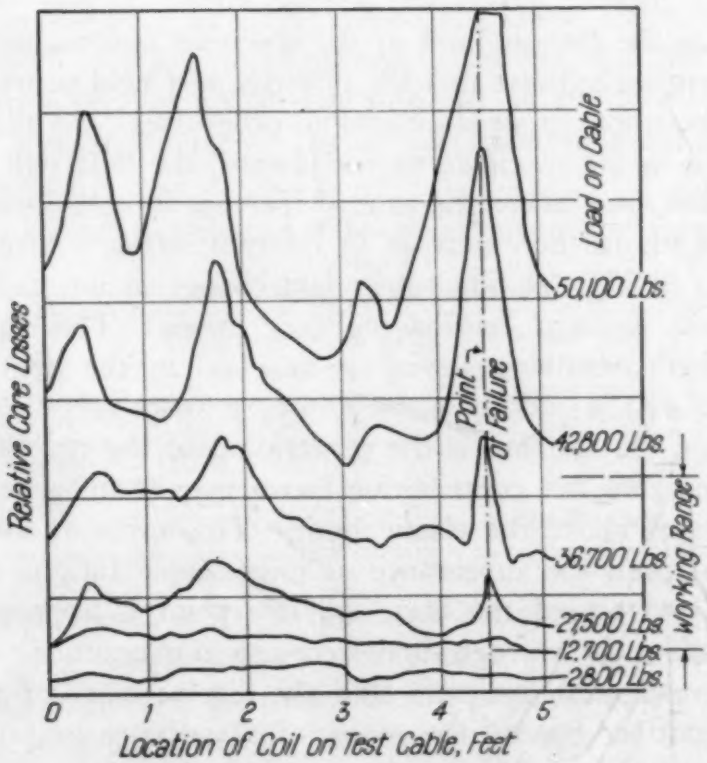

Fig. 22—Prediction of Point of Failure Under Load Test Frequency—4 KC. Cable—1.5-Inch Lang Lay.

increment of volume which is included in the zone under tension apparently reaches a definite value of internal stress before a further increment of volume is affected enough to influence the core losses.

The oscillations about the average curves for annealed specimens are not so pronounced. A surface layer under tension must first be established before regular oscillations begin.

Appearance of a fatigue crack has no affect whatever on the cyclograph reading.

Cable Testing

One useful practical application of fatigue prediction is the continuous inspection of mine hoist cable while in service. Magnetic means of accomplishing such an inspection have often been tried in the past with little success. The problem is certainly not simple to solve since a wire cable contains complex stresses and flux penetration cannot be calculated with even approximate accuracy. Effective penetration of a test field at a certain frequency will be much greater for a cable than for a solid bar of the same diameter.

Figs. 20, 21 and 22 illustrate some of the difficulties to be overcome if high frequency core loss determinations are used to inspect cable. On cable less than ¾ inch in diameter, no difficulty is experienced in detecting common faults which may give rise to fatigue failure (13). Cable 1 inch or more in diameter must be tested within a narrow range of test frequencies to get useful information which can be used to operate automatic recording and warning devices. For the cyclograph to operate such circuits, it is desirable to have all defects give an increase in core losses. The choice of proper test frequency is quite critical for some materials and cable sizes.

Cutting one wire, corroding outside wires or filing or abrading material from the surface of a cable throws higher than average stresses on the inner wires at this point. In order to detect the abnormally high stresses in the inner wires, the test flux must penetrate far enough to include them in the test field. Hammering the cable will redistribute the increased stresses more uniformly among the remaining load carrying wires as will service vibration. The higher level of unit stresses in the wire in the neighborhood of the defect still allows its detection.

Bruises or interior corrosion in a cable are much more easily detected than any other fault.

Recordings taken on a cable in service day after day will show a standard pattern of core loss variation over the length of the cable. If the core losses increase at some one point and continue to increase relative to the core losses in the rest of the cable as time goes on, the cable will fail in fatigue sooner or later at this point.

Conclusions

A high frequency oscillator which will compare total core losses in metal specimens seems admirably suited to rapid nondestructive stress comparisons. The method is only applicable in certain specific types of investigations and has definite limitations.

If used intelligently, core loss correlations can give useful information about stress conditions in ferrous and nonferrous metals. The possibility of predicting fatigue failure seems excellent in some instances.

The author would like to express his gratitude for the assistance received from the staff of the Materials Test Division of DuMont Laboratories in conducting these tests, to Mr. O. W. Ellis of the Ontario Research Foundation for his aid and to the metallurgical staffs of R.P.I. and the several manufacturers without whose collaboration this investigation could not have been carried out.

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DISCUSSION

Written Discussion: By J. K. Stanley, research engineer, Research Laboratories, Westinghouse Electric Corp., East Pittsburgh, Pa.

Testing with the cyclograph, for example in production, appears to be attractive and practical because of its rapidity and nondestructive nature. The cyclograph, it seems to the writer, could also be developed for research work. Has any attempt been made to calibrate the apparatus for use as a research tool rather than as a comparison instrument?

A research instrument for determining stresses in nonmagnetic materials would be a useful tool. For magnetic materials the direct determination of stresses is not easily realized, but for nonmagnetic metals the case is attractive. In magnetic materials the total losses depend upon the hysteresis and the eddy currents, and the hysteresis in turn depends not only on the stress but also on the grain size. If there were no hysteresis, as in nonmagnetic materials, and hence no grain size effect, the change in eddy currents in nonmagnetic material caused by a change in the resistivity would be a function of the internal stress. Such an instrument could very easily be useful in measuring stresses in rolled nonferrous materials.

Two questions occur to the writer: (a) How does one arrive at the frequency to use for a test? Is this done by trial and error? Is it not desirable to operate at the lowest possible frequency to minimize the "skin effect"? (b) Would it not be possible to determine the depth of a carburized case by studying the "skin effect" as a function of the frequency?

Author's Reply

We have always regarded the cyclograph as a comparative device and certainly not as a measuring instrument. It can be calibrated for a particular problem and particular type of steel and certain set of conditions so that it might be called a measuring device in that instance, but it must be recalibrated for each separate problem. Used as a research tool, it is quite feasible to follow this procedure. In production it is also practical if there is enough production to warrant the calibration on each problem.

As Mr. Stanley says, the determination of stresses in nonmagnetic materials is much simpler than in magnetic materials. If we could convince more people that they should use nonferrous materials for this reason, we would be very happy, but I do not think that is likely to be the case.

As Mr. Stanley mentions, the extraneous variables such as grain size do

cause trouble in operating on magnetic materials. Success in correlating with stresses depends on success obtained in eliminating such extraneous variables by preliminary testing or other methods mentioned. The specific questions as to test frequencies are answered in the affirmative. The best test frequency is chosen by trail and error. The approximate frequency that would be desirable to use can sometimes be guessed within a few thousand cycles, but actually there is an optimum test frequency which is governed by the balance between magnetic and electrical losses which make up the total core losses and the correlation between the physical property it is desired to evaluate and the magnetic and electrical properties being examined. In general, if you are not interested in the skin of the material you do not use high frequencies, but the best frequency is not necessarily the lowest.

The question as to correlating with case depth may also be answered in the affirmative. This instrument is used in production, and has been for some time, to compare case depth. This inspection is only possible over narrow limits of case depth. Pieces cannot be compared for case depth if the range is considerable, say from 0.010 to 0.090 inch or from 0.010 to 0.050 inch. A variation of double a case depth of 0.020 inch is within the range of the instrument. The practical range of comparison is governed by flux penetration, which is affected by the properties of the surface layers of material. The penetration of the field should coincide approximately with case depth to get satisfactory correlation with case depth, and if the case increases greatly in depth, flux penetration no longer coincides closely with case penetration.

METALLURGICAL CHARACTERISTICS OF INDUCTION-HARDENED STEEL

By James W. Poynter

Abstract

Specimens of S.A.E. 4340 steel, heat treated by induced high frequency (355,000-cycle) electric currents and quenched, have the same metallurgical characteristics as furnace-heated and quenched specimens. The depth of hardening is increased by increasing heating times (lower power input) to produce the same surface temperature or by heating to higher surface temperatures with the same power input. The transition between the hardened zone and the unaffected metal may be cone-shaped. The hardness of this transition zone is naturally less than that of the hardened zone and in some cases may be less than that of the unaffected metal.

Samples containing small carbide particles respond more readily to heat treatment than those in which the carbide particles are larger.

No evidence is found to indicate that induction heating results in more rapid solution and transformation rates and in the absence of grain coarsening at higher temperatures. It is believed that the effect of frequency on depth of hardening has been overemphasized since the rate of heat flow also has a definite effect on the depth of hardening.

While the evidence obtained regarding the so-called "superhardness" is not conclusive, it seems likely that the increased hardness is related to the presence of internal stresses and not to any difference in the metallurgy.

MUCH has been written on the subject of induction hardening, but only a relatively few papers contain actual data on the metallurgical characteristics and behavior of induction-heated steels. It is hoped that this present work will extend and add to the existing information.

This investigation is conducted to determine the metallurgical characteristics of steel hardened by induced high frequency electric

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currents and to determine if these characteristics differ from those of steel furnace-heated and quenched in the usual way. Included are studies on the effect of various power inputs on the depth and length of the hardened zone produced, on the shape of this hardened zone in a cylindrical specimen, on the metallographic structure, and on the so-called "superhardness".

Experimental Procedure

A coil 1¼ inch I. D. and approximately 2⅝ inches high was made from ¼-inch copper tubing. The coil contained 5¾ complete turns with a ⅛-inch space between turns. Power was supplied by a 30-kilowatt spark gap-type converter.

The specimens, 1 inch in diameter by 4½ inches long, were supported vertically inside the coil by a transite plate ½ inch below the coil. The test piece was centered in the coil manually within plus or minus $\frac{1}{32}$ inch. The temperature was measured with a disappearing filament-type optical pyrometer focused on the specimen between the center turns of the coil. The heating time was measured by a stop watch.

At the conclusion of the heating cycle, the specimen was lifted out of the coil by tongs and agitated in the quenching medium. This transfer was made in 3 seconds or less. The vast majority of the specimens were oil-quenched, a few, as noted in the data or discussion, were water-quenched.

Any desired surface temperature was held by successive, short applications of power after the surface comes up to temperature. The frequency and duration of these applications were controlled manually so that the temperature shown by the optical pyrometer did not fluctuate appreciably.

A disk ⅜ inch thick for transverse hardness survey and metallographic examination was taken from the center of the heated zone of each specimen. The top of the specimen was split longitudinally and half of it was ground smooth and etched with an aqueous solution containing 50 per cent concentrated hydrochloric acid to show the pattern produced by the hardened zone.

The disks were ground flat on both sides for the transverse hardness survey. Rockwell C readings were obtained every $\frac{1}{16}$ inch along a diameter, the first reading being taken $\frac{1}{16}$ inch from the edge. To obtain additional points, the disks were turned 90 degrees and a set

of readings taken along this diameter. In this second set, however, the first reading was taken $\frac{1}{32}$ inch from the edge. This gives a series of hardness values for every $\frac{1}{32}$ inch from surface to center.

When longitudinal hardness surveys were made, about 0.005 inch was ground off two surfaces 180 degrees apart to insure readings unaffected by the very thin surface oxide layer.

The samples were taken from two lots of annealed S.A.E. 4340 electric furnace steel of very similar compositions (Table I). Except for the few samples induction-hardened in the "as-received" condi-

Table I

Designation	A	B
Producer	Crucible	Republic
Melt No.	601327	616339
Carbon	0.40%	0.40%
Manganese	0.67	0.65
Phosphorus	0.012	0.019
Sulphur	0.016	0.016
Nickel	1.83	1.76
Chromium	0.78	0.72
Molybdenum	0.25	0.26

Both are electric furnace steels.

tion, all the test pieces were normalized at 1600 degrees Fahr. (870 degrees Cent.) for 1 hour and air-cooled. Some of the normalized bars were tempered at 1050 degrees Fahr. (565 degrees Cent.) for 1 hour, some were hardened by oil quenching from 1525 degrees Fahr. (830 degrees Cent.), and others were spheroidized. The spheroidizing treatment consisted of 16 hours at 1300 degrees Fahr. (705 degrees Cent.), 2½ hours at 1425 degrees Fahr. (775 degrees Cent.), 2 hours at 1300 degrees Fahr. (705 degrees Cent.), and furnace cooling.

Except for the specimens which were fully hardened, all specimens were heat treated in $1\frac{1}{16}$ inch or larger size and subsequently turned down to 1 inch to remove all decarburized metal. Because of the high resulting hardness of the heat treated specimens, it was necessary to turn them to 1 inch before furnace hardening. On these samples, the data obtained near the surface were disregarded, since there was a possibility of some decarburization.

Results

Depth of Hardening—Tests were made on normalized and drawn specimens to determine the reproducibility of the cross-sec-

tional hardening pattern under the same condition. It was found that the patterns obtained were quite reproducible for the same power setting and the same surface temperature, even though the heating time varied somewhat. The results shown in Fig. 1 for a power

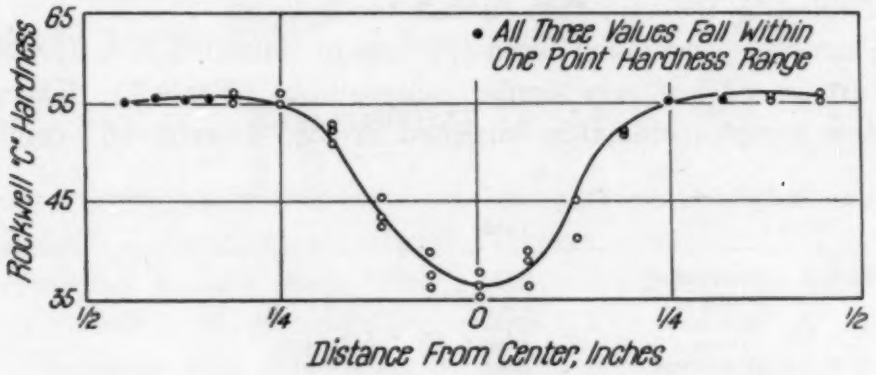

Fig. 1—Reproducibility of Cross-Sectional Hardening Pattern Under Same Operating Conditions. Power input setting 15 kilowatts. Surface temperature 1500 degrees Fahr. (815 degrees Cent.). Specimens normalized and tempered.

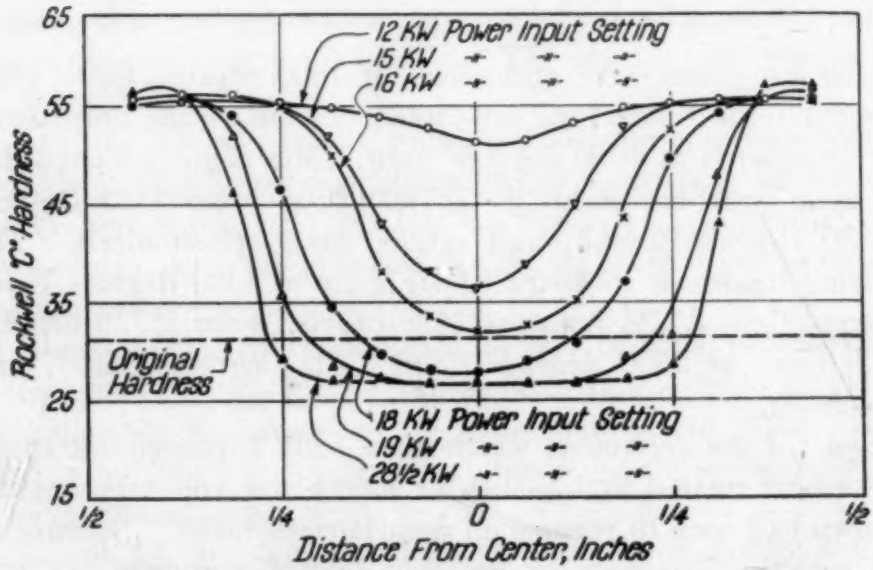

Fig. 2—Effect of Power Input on Depth of Hardening. Surface temperature 1500 degrees Fahr. (815 degrees Cent.). Specimens normalized and tempered.

input of approximately 15 kilowatts are typical of the reproducibility obtained.

In Figs. 1 through 10, the Rockwell "C" hardness values at $\frac{1}{16}$-inch intervals are plotted against the distance from the center of the disk along a diameter. Values at $\frac{1}{32}$-inch intervals are plotted where this is necessary to show the shape of the curve.

The greater the power input, the shallower was the depth of hardening produced on the normalized and drawn specimens (Fig. 2) when the surface was heated to 1500 degrees Fahr. (815 degrees Cent.). In all cases, the surface hardness was Rockwell C-55 to 56.

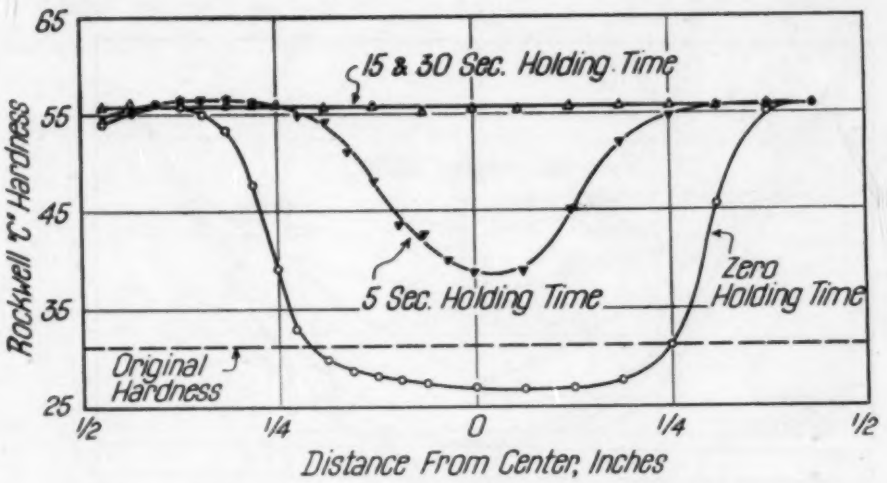

Fig. 3—Effect of Time at 1500 Degrees Fahr. (815 Degrees Cent.) on Depth of Hardening. Power input setting 19 kilowatts. Specimens normalized and tempered.

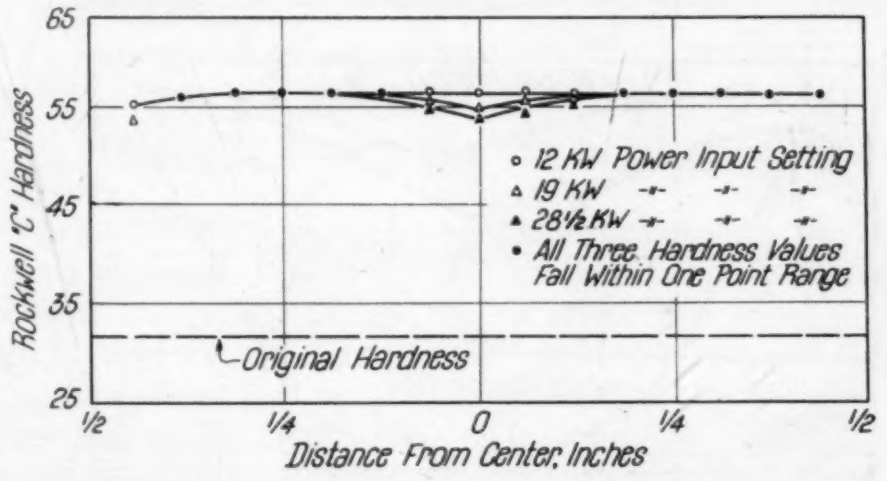

Fig. 4—Effect of Power Input on Depth of Hardening. Surface temperature 1600 degrees Fahr. (870 degrees Cent.). Specimens normalized and tempered.

The depth of hardening varied from slightly over ⅛ inch with a power input of 28½ kilowatts to almost complete cross-sectional hardening with a power setting of 12 kilowatts. The times varied from 22 seconds to 52.5 seconds, the shorter heating time being associated with the higher power input.

The depth of hardening increased rapidly when normalized and

drawn specimens were held at a surface temperature of 1500 degrees Fahr. (815 degrees Cent.) (Fig. 3). Under the condition of test, times of 15 seconds at temperature were sufficient to harden through the cross section.

When heated to 1600 degrees Fahr. (870 degrees Cent.) any

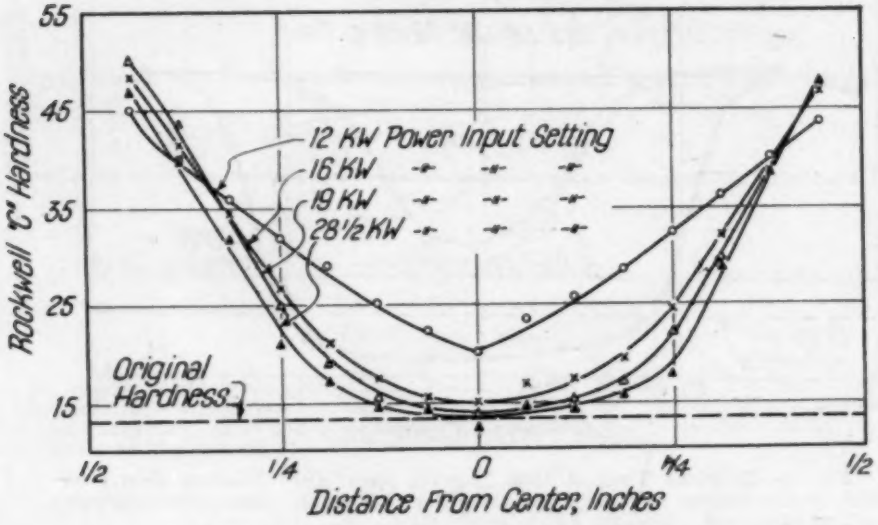

Fig. 5—Effect of Power Input on Depth of Hardening. Surface temperature 1500 degrees Fahr. (815 degrees Cent.). Specimens spheroidized.

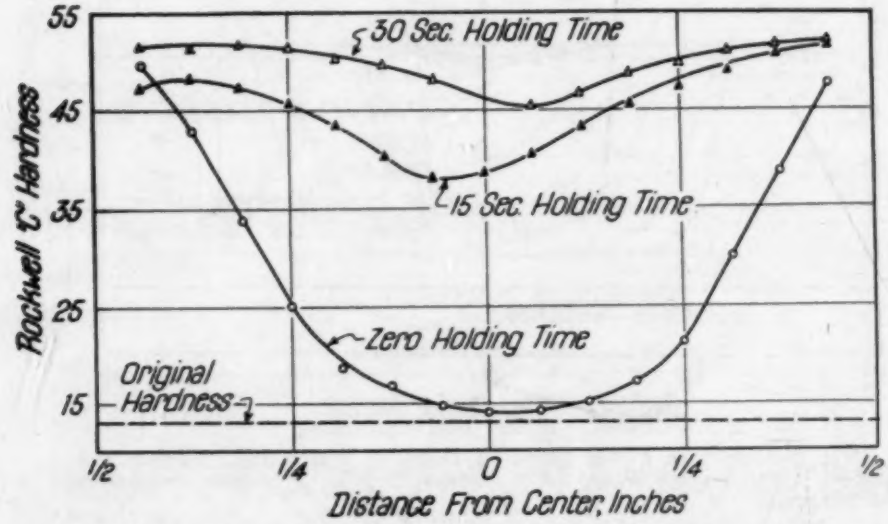

Fig. 6—Effect of Time at 1500 Degrees Fahr. (815 Degrees Cent.) on Depth of Hardening. Power input setting 19 kilowatts. Specimens spheroidized.

power input tried (12 to 28½ kilowatts) was sufficient to completely harden the cross section of the normalized and drawn test specimens (Fig. 4). The only difference noted was that the specimen heated most rapidly (24.3 seconds) had a center hardness of 2 Rockwell C

less than the sample heated the least rapidly (66.2 seconds). This variation is close to the limit of accuracy of the hardness tester and consequently is not considered to be of great significance.

The spheroidized samples did not respond adequately when the surface temperature was raised to 1500 degrees Fahr. (815 degrees Cent.) with various power inputs (Fig. 5). The maximum hardness found $\frac{1}{16}$ inch from the surface was Rockwell C-50, as against the

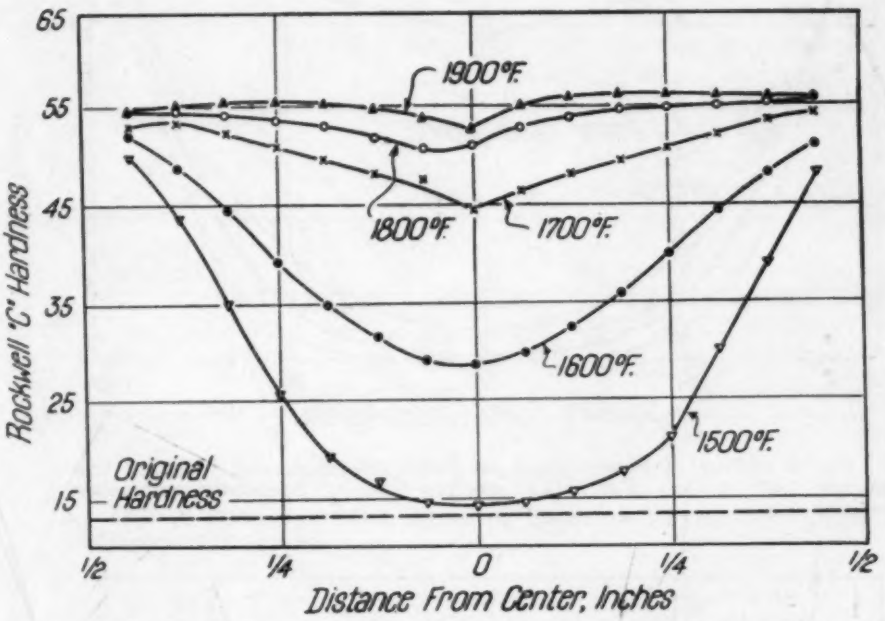

Fig. 7—Effect of Surface Temperature on Depth of Hardening of Spheroidized Specimens. Power input setting 19 kilowatts.

Rockwell C-55 of the normalized and drawn samples. The hardness dropped off rather rapidly toward the center of the bars and this drop was not greatly changed by variations in the power setting. The 12-kilowatt power setting, which produced almost full hardening of the normalized and tempered specimens, produced only slightly greater hardening than that given by the other power settings.

Hardening was obtained when the surface temperature of the spheroidized specimen was held at 1500 degrees Fahr. (815 degrees Cent.) for 30 seconds (Fig. 6). Partial hardening was obtained with 15 seconds at temperature but neither time was sufficient to produce the maximum hardness of which this steel was capable.

It was found necessary to raise the surface temperature to 1800 degrees Fahr. (980 degrees Cent.) or more (with a power input of 19 kilowatts) in order to completely harden the spheroidized specimen $\frac{1}{16}$ inch from the surface (Fig. 7). Under this condition, hardening

was obtained throughout the cross section of the piece. When the surface was heated to 1700 degrees Fahr. (925 degrees Cent.), slightly lower hardness was obtained $\frac{1}{16}$ inch from the surface and the hardness dropped off toward the center.

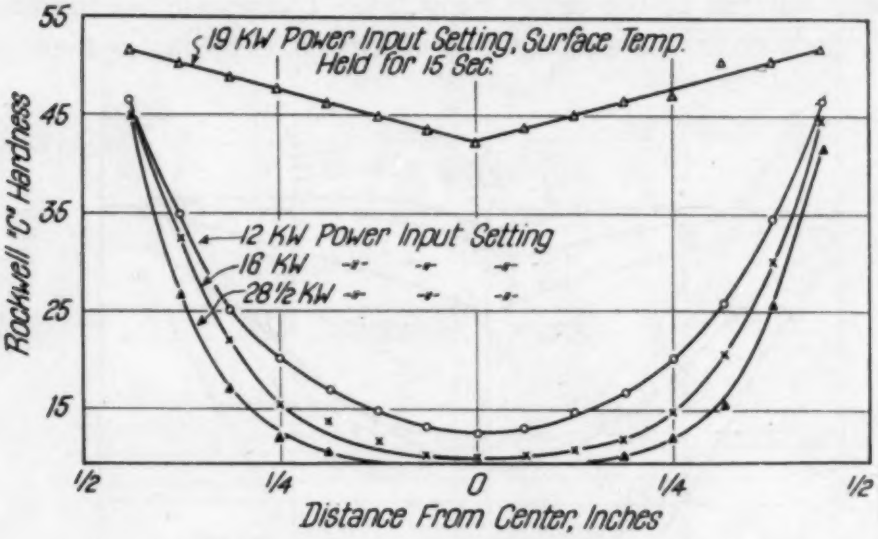

Fig. 8—Effect of Power Input on Depth of Hardening. Surface temperature 1500 degrees Fahr. (815 degrees Cent.). Specimens as-received condition.

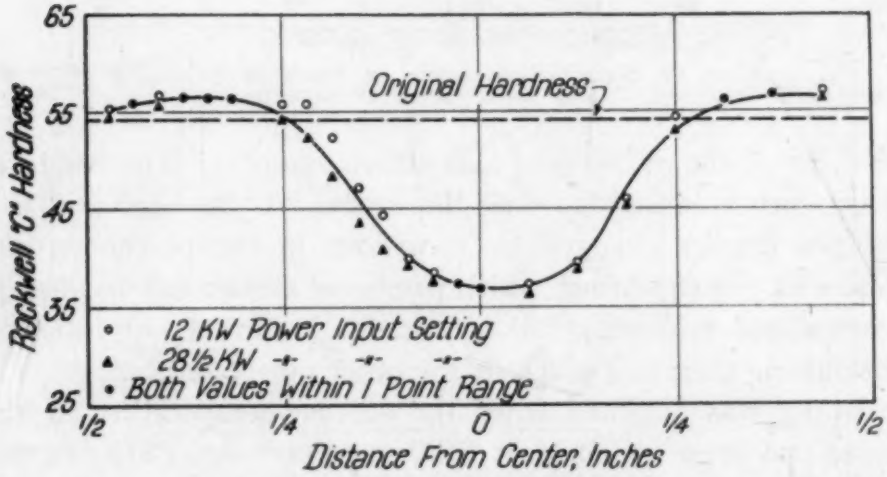

Fig. 9—Effect of Power Input on Depth of Hardening. Surface temperature 1500 degrees Fahr. (815 degrees Cent.). Specimens fully hardened.

The specimens of the "as-received" material (apparently annealed) responded to induction heating to a surface temperature of 1500 degrees Fahr. (815 degrees Cent.) in a similar manner to the spheroidized samples (Fig. 8).

When the specimens were fully hardened before the induction heating, an entirely different behavior was found (Fig. 9). A zone under the surface (¼ inch under the test conditions used) is rehardened and the zone nearer the center is tempered. The hardness values and pattern obtained were not appreciably affected by variations in the power used.

The results obtained on the limited number of hardened and

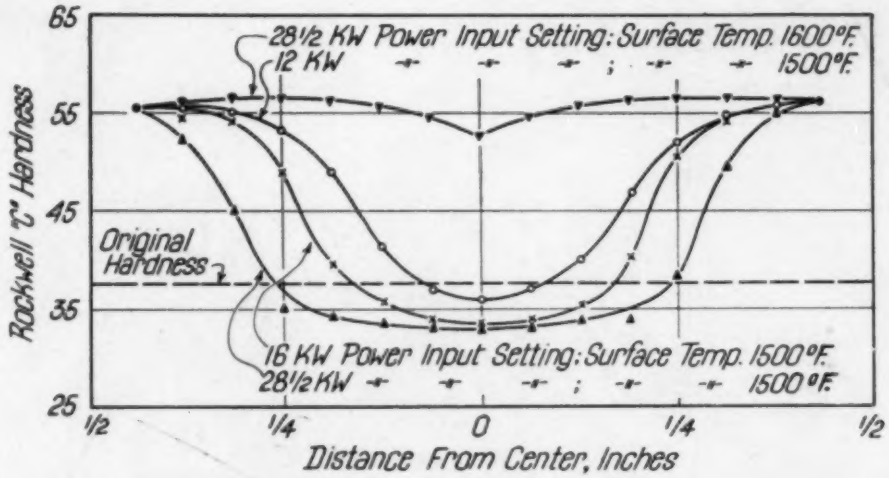

Fig. 10—Effect of Power Input on Depth of Hardening. Specimens quenched and tempered.

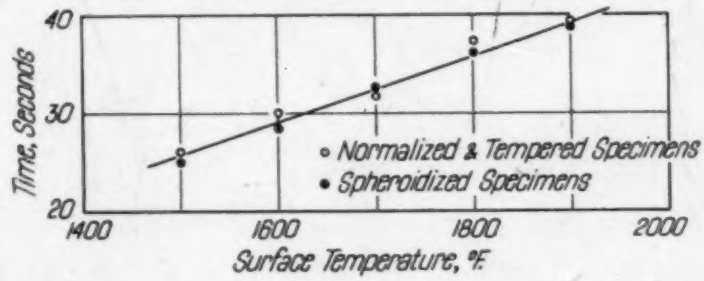

Fig. 11—Time Required to Heat to Various Surface Temperatures. Power input setting 19 kilowatts.

drawn specimens tested (Fig. 10) were very similar to those obtained on the normalized and tempered bars.

With a power input of 19 kilowatts, and temperatures above the critical, the time to reach a given surface temperature is a linear function of the temperature (Fig. 11). No significant difference in time was noted between two of the extremes in structure (spheroidized and normalized and drawn).

Longitudinal Extent of Hardening—No perceptible difference in extent of the hardened zone was noted for the normalized and

tempered specimens heated to 1500 degrees Fahr. (815 degrees Cent.) with various inputs (Fig. 12). At the end of the hardened zone, the hardness dropped off very sharply to a hardness of 3 to 4 points Rockwell C lower than the original hardness of the material. In the next ¼ inch, the hardness values increased to the original value.

The length of the hardened zone was not increased until the surface temperature was held at 1500 degrees Fahr. (815 degrees Cent.) for 30 seconds. Under these conditions and with a power

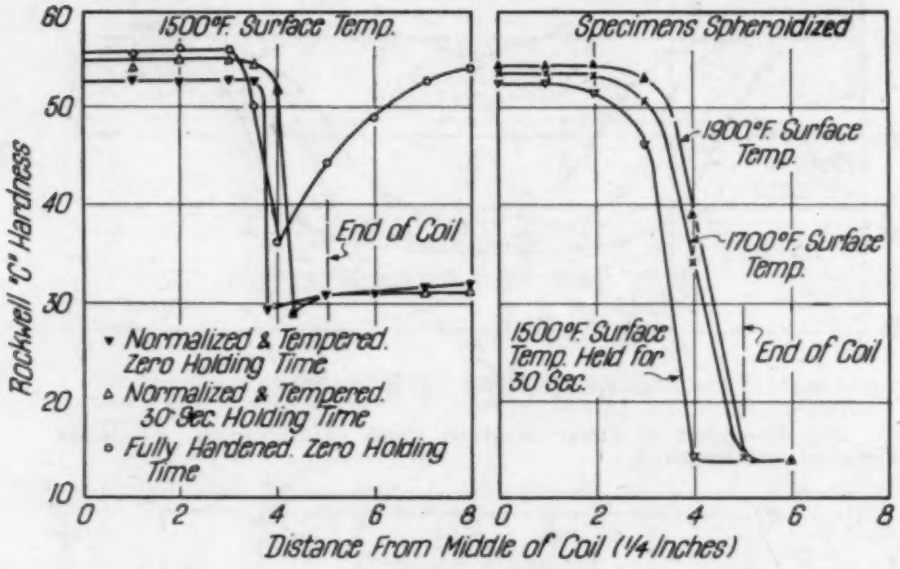

Fig 12—Longitudinal Extent of Hardening. Power input setting 19 kilowatts.

input of 19 kilowatts, the length was increased less than $\frac{1}{16}$ inch.

A very similar behavior was found for the spheroidized samples. In these specimens, however, the hardness began to decrease gradually about ¼ inch before the sudden drop and did not drop below the original hardness. It should be noted, however, that the original hardness of the spheroidized material was appreciably lower than that of the normalized and quenched material.

The behavior of the induction-rehardened specimens in the full-hard condition did not differ from the expected behavior. As with specimens in the other conditions, the hardness dropped off very sharply ¼ to ⅜ inch short of the coil end. The hardness then gradually increased with distance from the heated zone until it reached the original value.

Contour of Hardened Metal—A study was made of the changes

produced in the longitudinal contour of the hardened metal by the variations in power input, time, surface temperature and previous structure. In all cases, the extent of the hardened zone along the

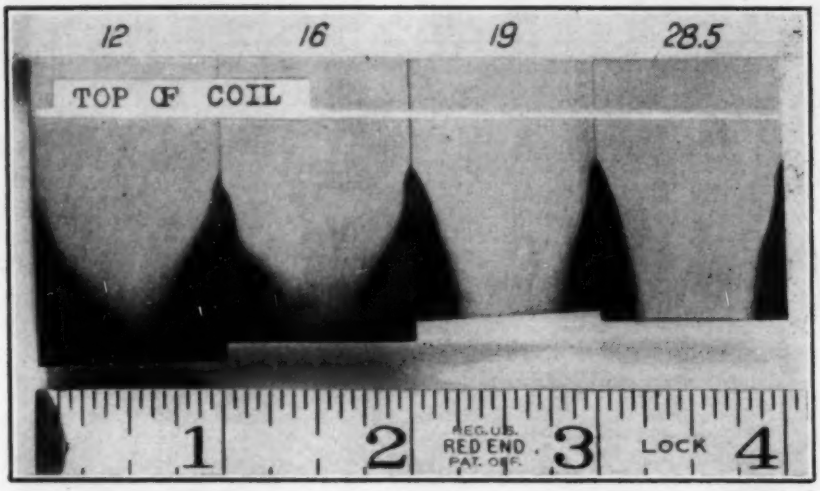

Fig. 13a—Effect of Increased Power Input to Produce the Same Surface Temperature (Approximately 1500 Degrees Fahr.). Specimens normalized and tempered. Figures above specimens show power input in kilowatts.

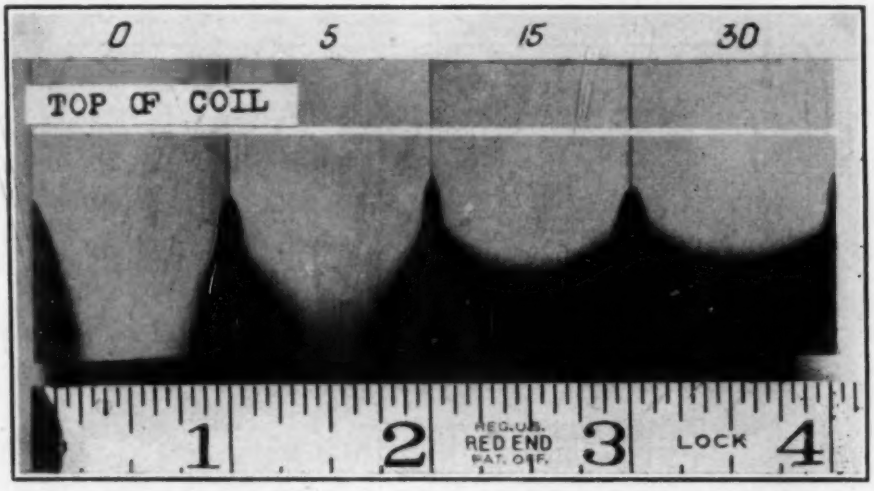

Fig. 13b—Effect of Time at a Surface Temperature of Approximately 1500 Degrees Fahr. (815 Degrees Cent.). Specimens normalized and tempered. Power input 19 kilowatts. Figures above specimens show holding time in seconds.

surface was unaffected by changes in these variables (Figs. 13, 14 and 15). This conformed to the results obtained by the longitudinal hardness survey previously reported.

At 1500 degree Fahr. (815 degree Cent.) surface temperature, the depth of hardening in the middle of the specimens decreased with

increasing power input (Figs. 13a and 15a). With the lowest power input, 12 kilowatts, substantially through hardening in the middle resulted. Surface hardening only was produced with 28½-kilowatt input. Similar changes in the hardening pattern occurred with a 1600

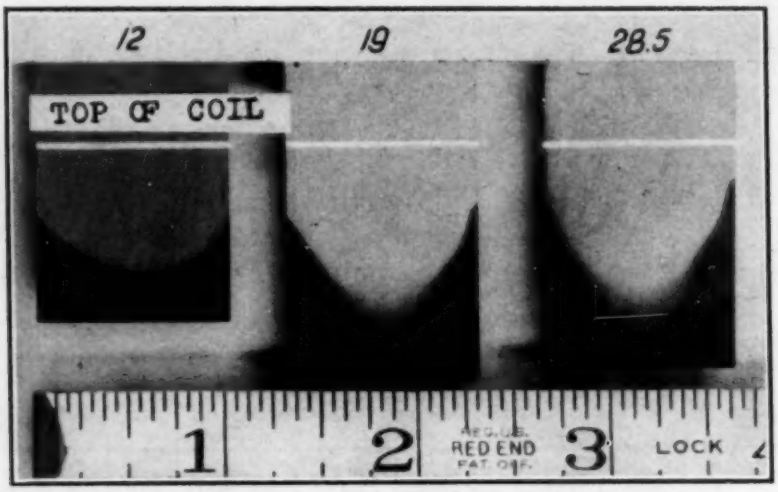

Fig. 13c—Effect of Increased Power Input to Produce the Same Surface Temperature (Approximately 1600 Degrees Fahr.). Specimens normalized and tempered. Figures above specimens show power input in kilowatts.

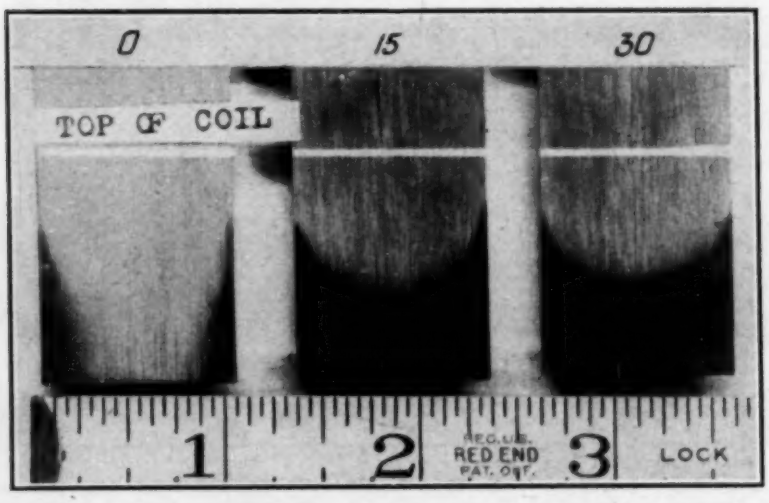

Fig. 13d—Effect of Time at a Surface Temperature of Approximately 1500 Degrees Fahr. (815 Degrees Cent.). Specimens spheroidized. Power input 19 kilowatts. Figures above specimens show holding time in seconds.

degree Fahr. (870 degree Cent.) surface temperature although the amount of hardening was so greatly increased that through hardening of the middle of the specimens was obtained in all cases (Fig. 13c).

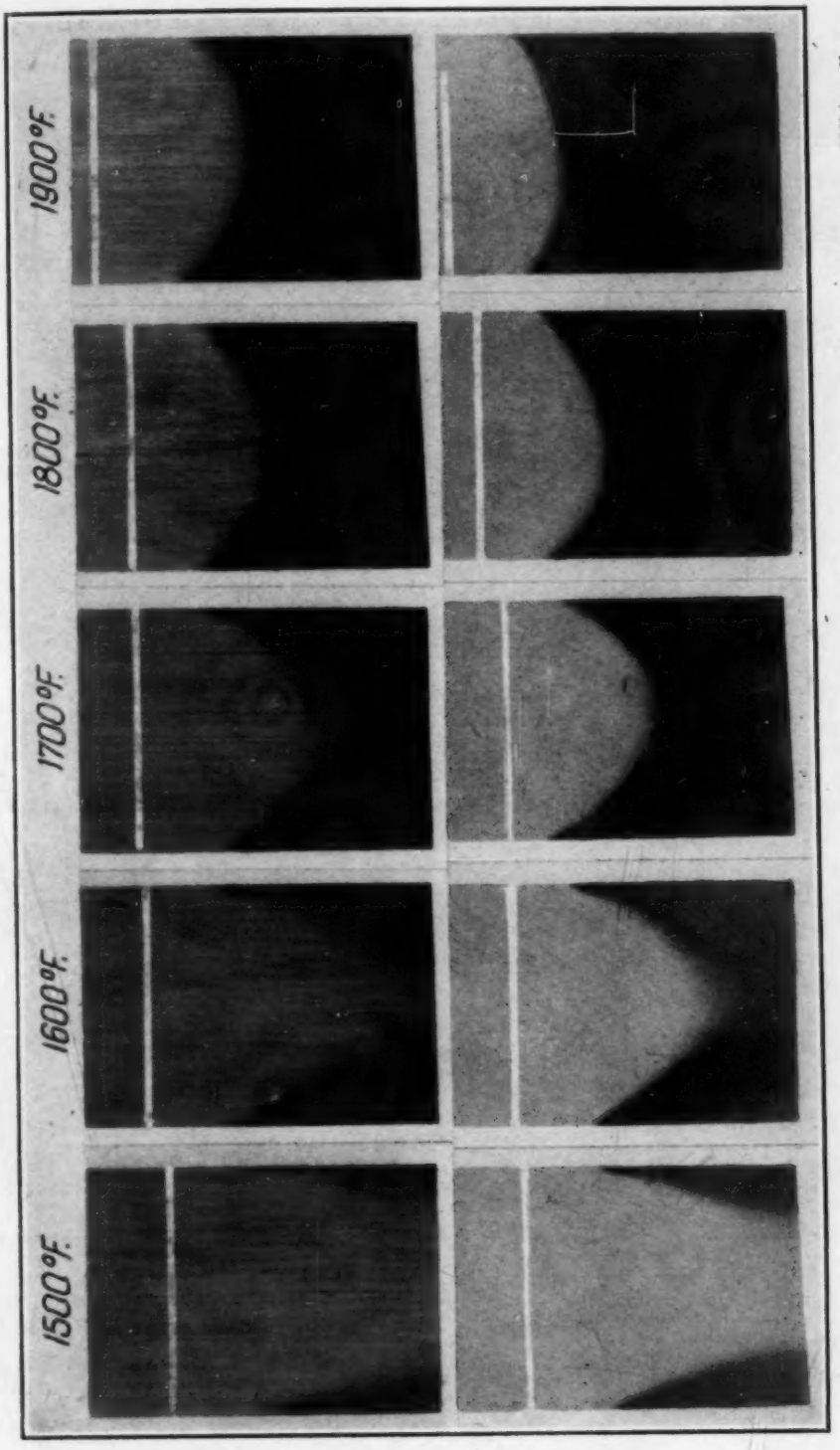

Fig. 14—Effect of Surface Temperature on Longitudinal Hardness Pattern. Power input 19 kilowatts. White line shows top of coil. Upper specimens spheroidized. Lower specimens normalized and drawn at temperatures shown at top.

Increases in the time at which the sample was held at a definite surface temperature, 1500 degrees Fahr. (815 degrees Cent.), produced increases in the volume of metal hardened (Figs. 13b and 13d). Holding times of 15 seconds were necessary for the shape of

the longitudinal hardening patterns to even approach the rectangular. Very little difference in hardening pattern was noted between the specimens held for 15 and for 30 seconds.

Increases in the surface temperature also produced increases in the volume of metal hardened (Fig. 14). The hardness pattern pro-

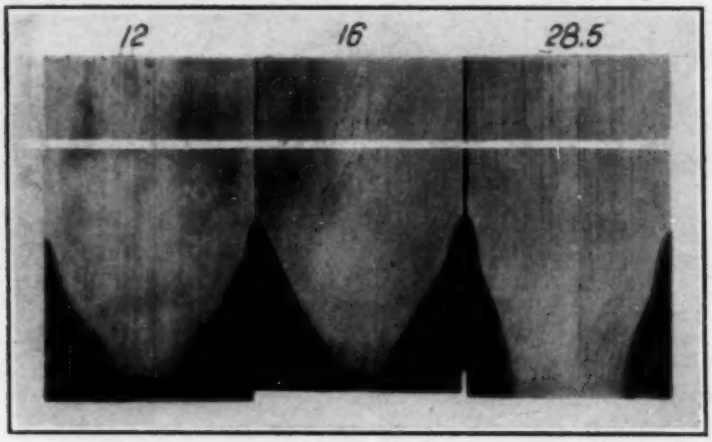

Fig. 15a—Effect of Increased Power Input to Produce the Same Surface Temperature (Approximately 1500 Degrees Fahr.). Specimens quenched and tempered. Figures above specimens show power input in kilowatts.

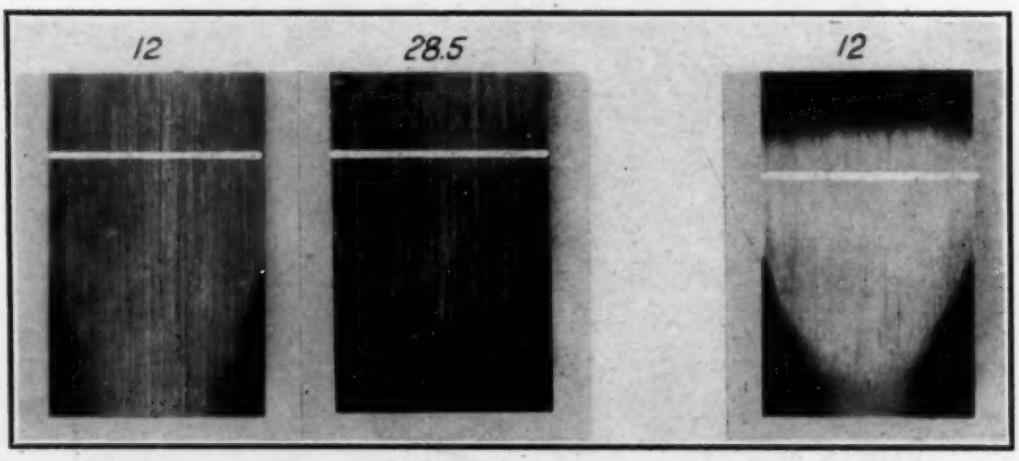

Fig. 15b (The Two Specimens on the Left Constitute Fig. 15b.)—Effect of Increased Power to Produce the Same Surface Temperature (Approximately 1500 Degrees Fahr.). Specimens "as-received" condition. Figures above specimens show power input in kilowatts.

Fig. 15c (The Single Specimen on the Right Constitutes Fig. 15c.)—Effect of Induction Hardening Specimen Previously Quenched. Power input 12 kilowatts. Surface temperature 1500 degrees Fahr. (815 degrees Cent.).

duced by heating to a surface temperature of 1900 degrees Fahr. (1040 degrees Cent.) was similar in extent to those produced by holding for 30 seconds at a surface temperature of 1500 degrees Fahr. (815 degrees Cent.). However, it will be shown that a surface temperature of 1900 degrees Fahr. (1040 degrees Cent.) pro-

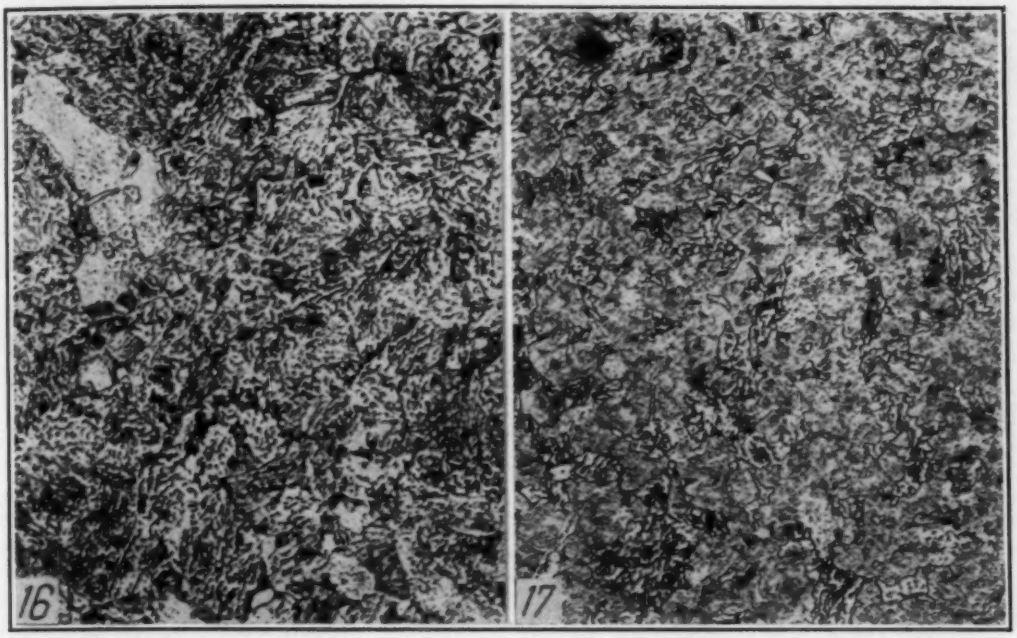

Fig. 16—Structure of Normalized and Drawn Steel Before Induction Heating. Nital etch. × 500.
Fig. 17—Partially Transformed Structure Found When Specimen is Heated to a Surface Temperature of 1400 Degrees Fahr. (760 Degrees Cent.). Nital etch. × 500.

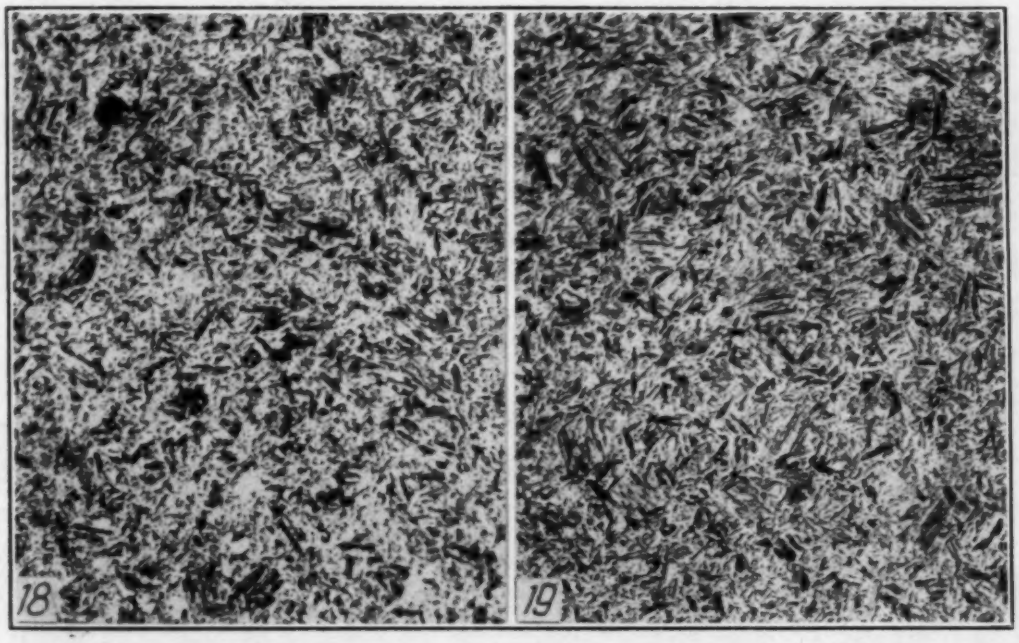

Fig. 18—Martensitic Structure Found When Normalized and Tempered Specimen is Heated to a Surface Temperature of 1500 Degrees Fahr. (815 Degrees Cent.). Nital etch. × 500.
Fig. 19—Slight Coarsening of Martensitic Structure When Specimen is Heated to a Surface Temperature of 1700 Degrees Fahr. (925 Degrees Cent.). Nital etch. × 500.

duces a very coarse grain structure near the surface.

The same inverted cone-shaped line of separation was found in

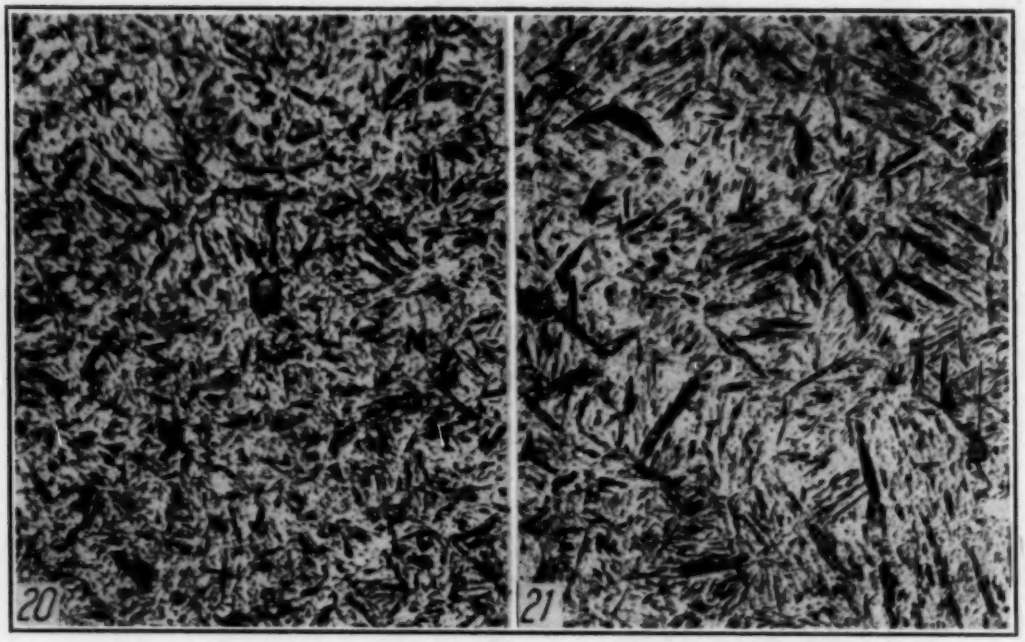

Fig. 20—Coarsening of Martensitic Structure When Specimen is Heated to a Surface Temperature of 1800 Degrees Fahr. (980 Degrees Cent.). Nital etch. × 500.
Fig. 21—Pronounced Coarsening of Martensitic Structure When Specimen is Heated to Surface Temperature of 1900 Degrees Fahr. (1040 Degrees Cent.). Nital etch. × 500.

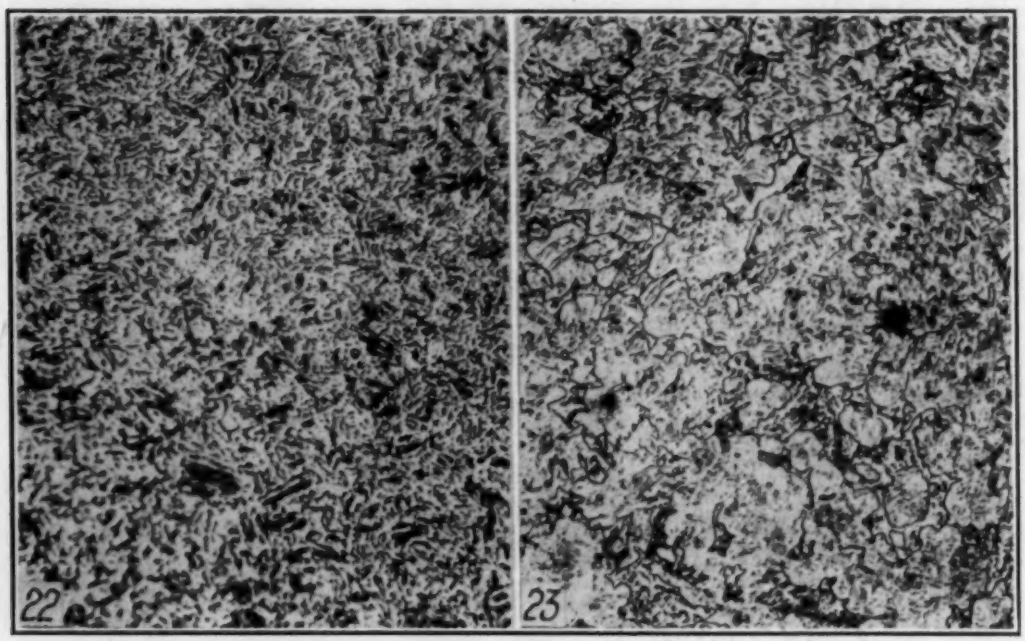

Fig. 22—Beginning of Transition Between Martensitic Surface Zone and Unaffected Center; Characterized by Presence of Small Undissolved Carbide Particles. Nital etch. × 500.
Fig. 23—The End of the Transition Between the Surface Zone and Unaffected Center; Characterized by Clusters of Carbides or Partially Spheroidized Pearlite in a Matrix Similar to a Low Carbon Martensite. Nital etch. × 500.

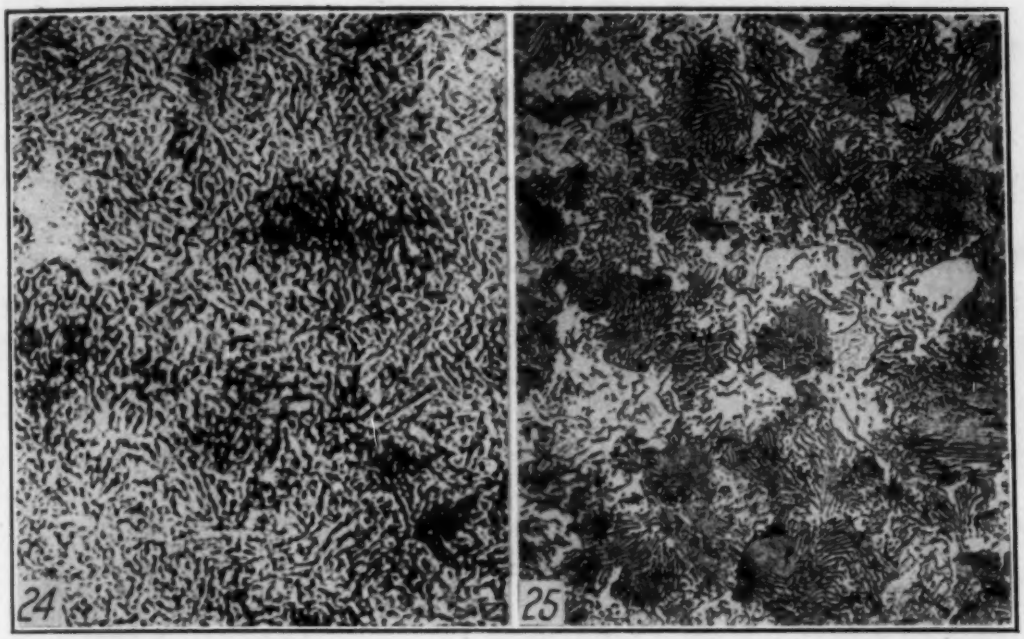

Fig. 24—Same as Fig. 23, at a Higher Magnification. 2 per cent picral etch. × 1300.
Fig. 25—Pearlitic Structure of Steel in the As-Received Condition. Nital etch. × 500.

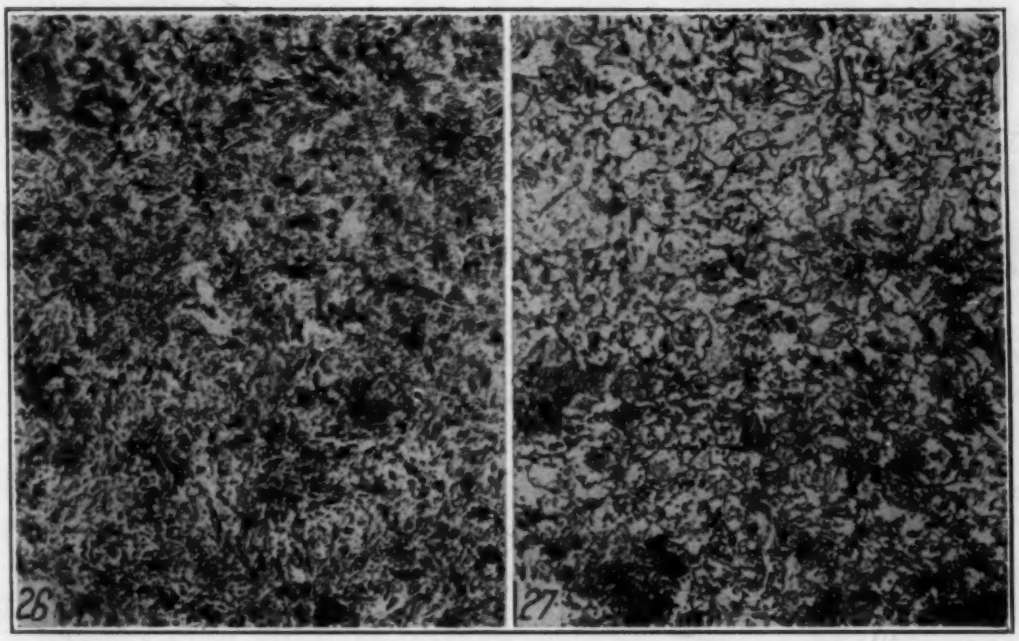

Fig. 26—Type of Martensitic Structure Obtained When Specimens in "As-Received" Condition Are Heated to a Surface Temperature of 1500 Degrees Fahr. (815 Degrees Cent.). Nital etch. × 500.
Fig. 27—Structure Resulting from Spheroidizing Treatment. Nital etch. × 500.

the induction-heated sample previously fully hardened as was found in the specimens induction-heated from the normalized and tempered condition (Fig. 15c). The pattern obtained in this case was char-

acterized by the fact that the soft light-colored zone was produced by tempering of the original martensitic structure and did not represent the original structure. The line of demarcation between this zone and the untempered metal at the end of the specimens was a straight line perpendicular to the surface.

In all of the specimens examined, the line of demarcation between the hardened and the soft zone varied in sharpness with the previous structure of the specimens. A rather gradual separation

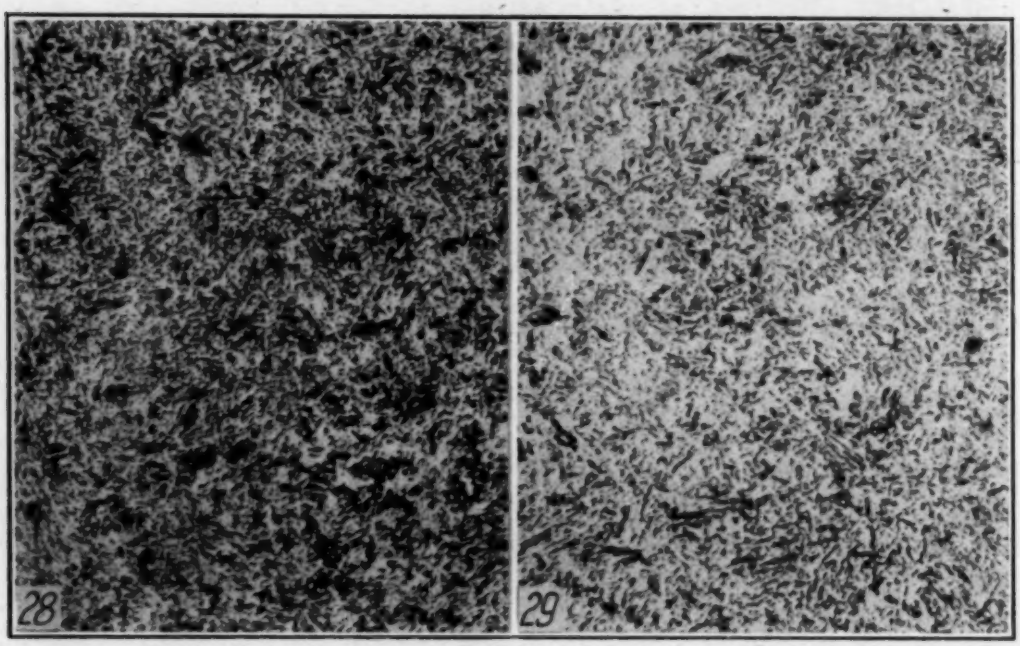

Fig. 28—Surface Structure of Spheroidized Specimens Heated to a Temperature of 1500 Degrees Fahr. (815 Degrees Cent.); Characterized by Carbide Particles in a Poorly Defined Martensitic Matrix. Nital etch. × 500.

Fig. 29—Carbide Clusters in Transition Zone of Spheroidized Specimens. Nital etch. × 500.

appeared in the spheroidized specimens while sharp separations were noted on the quenched, quenched and tempered, or normalized and tempered specimens. This difference was clearly shown in Fig. 14.

Metallographic Examination—The preliminary normalizing and tempering heat treatment of the specimens produced a structure resembling a poorly defined pearlite or a well-tempered martensite (Fig. 16). When induction-heated to an apparent surface temperature of 1400 degrees Fahr. (760 degrees Cent.), the carbides coalesced and some tended to go into solution (Fig. 17). This structure had a lower hardness than the original structure. At surface temperatures of 1500 degrees Fahr. (815 degrees Cent.) or higher, a

uniform, well defined martensitic structure was found in the hardened surface zone (Fig. 18). No carbides were noted. The structure tended to coarsen as the surface temperature increased. The coarsening was noticeable at 1700 degrees Fahr. (925 degrees Cent.) (Fig. 19) and very pronounced at 1800 and 1900 degrees Fahr. (980 and 1040 degrees Cent.) (Figs. 20 and 21).

The beginning of the transition zone between the fully hardened surface and the unchanged core was characterized by the appearance of undissolved carbide particles (Fig. 22) in the martensitic matrix. The carbide particles became more numerous toward the center. The structure of the inner portion of the transition zone consisted of clusters of carbides or partially spheroidized pearlite in a matrix similar to a low carbon martensite (Figs. 23 and 24).

The relative amounts of the hardened zone, transition zone and core were affected by changes in the power input and time, but the type of structure in each zone remained unchanged. The lower power inputs, and consequently longer heating times, produced deeper hardening. Likewise, holding at a fixed surface temperature for increasing times deepened the hardened zone without affecting the type of structure found in each zone.

With the lowest power input (12 kilowatts) or with holding times of 15 seconds at 1500 degrees Fahr. (815 degrees Cent.), no transition zone or unaffected core was found. When a surface temperature of 1600 degrees Fahr. (870 degrees Cent.) or higher was reached, the transition and unaffected zones were not found with any power input used.

In the "as-received" condition the steel had a pearlitic structure (Fig. 25). The response in this condition on induction heating was not as rapid as that of the normalized and tempered specimens. Undissolved carbide particles were found near the surface even though a low power input and long heating times were used. The appearance of the martensite in the hardened zone was not as acicular as that found in specimens which were induction-heated from other conditions (Fig. 26). In other respects the structure of these specimens was the same as that previously described.

To determine the effect of the carbide particle size on behavior on induction heating, specimens in the spheroidized condition were prepared (Fig. 27). Sufficient solution of carbides to produce a fully-hardened martensitic structure did not take place with any power input used to produce a surface temperature of 1500 degrees

Fahr. (815 degrees Cent.) or by holding at 1500 degrees Fahr. (815 degrees Cent.) for 30 seconds. The surface structure of these specimens consisted of undissolved carbide particles in a matrix (Fig. 28). This matrix appeared to be martensitic, but was neither clearly defined needles nor nodular.

When the surface temperature was 1600 degrees Fahr. (870 degrees Cent.) and the power input 19 kilowatts, the surface zone was a relatively uniform fine martensite containing small scattered carbides. The number of carbides increased toward the center of the specimen. The structure of the transition zone was similar to the transition zone in the normalized and tempered specimen previously described. Near the end of this zone, clusters of carbides were noted (Fig. 29).

At a 1700 degree Fahr. (925 degree Cent.) surface temperature, the same structures were observed except that some evidence of carbide solution was found at the center of the specimen. At 1800 degrees Fahr. (980 degrees Cent.) the structures were similar. The appearance of the remaining carbide particles near the surface indicated that they had been rather large particles which had not had time to completely dissolve.

When the surface temperature was 1900 degrees Fahr. (1040 degrees Cent.), the first evidence of coarsening of the martensite in the spheroidized specimens was noted. No carbide particles were found in the outer half of this specimen.

The hardened zone of the specimen which was fully hardened before induction heating had the same microstructure as that of the normalized and tempered specimens induction-heated under the same conditions. The transition zone was also similar but gradually changed to a tempered structure in the center of the specimen.

For comparison, a specimen in the normalized and tempered condition was induction-heated and water-quenched. No difference was detected between the microscopic structures of this specimen and of the oil-quenched specimens.

Superhardness—For purposes of comparison, a few samples were induction-heated and water-quenched. Higher hardnesses were obtained near the surface and consequently a more detailed investigation was made.

A specimen was induction-heated and water-quenched. Flats were ground 180 degrees apart and hardness values taken at intervals on both surfaces. The bar was split longitudinally; cut trans-

Table II
Change in Hardness Found on Cutting Induction-Heated and Water-Quenched Bar

	Side A		Side B	
	Before Cutting	After Cutting	Before Cutting	After Cutting
	Rockwell C		Rockwell C	
	....	60.9	....	61.1
	62.1	55.2	62.6	59.1
	61.6	61.1	62.4	59.8
	62.6	60.5	62.1	60.4
	62.0	59.5	62.2	59.9
	62.5	60.0	62.2	61.5
	62.3	60.9	62.0	60.6
	63.0	61.5	62.9	59.0
	63.0	61.1	62.7	61.1
Average	62.4	60.1	62.4	60.3

Specimen A38.

versely at intervals; and a second set of hardness readings taken ⅛ inch from the first. The resulting decrease in hardness is given in Table II. When split, it was noted the bar halves bowed about 0.01 inch. Extreme care was used in cutting the bar with a water-cooled cut-off wheel to avoid any possible heating and tempering.

After a set of hardness readings was made on a bar prepared in a similar manner, the bar was drawn for an hour at 325 degrees Fahr. (165 degrees Cent.) and a second set of hardness readings taken. A third set of hardness readings was taken after the bar

Table III
Effect of Various Treatments on the Rockwell C Hardness of an Induction-Heated Specimen

	Induction-Heated Water-quenched	Same, plus 1 hour at 325 degrees Fahr. (165 degrees Cent.)	Same, plus 1 hour at 1525 degrees Fahr. (830 degrees Cent.) in furnace Water-quenched
	63	59	57
	62.5	60	57
	61.8	58.5	57
	62.8	58.5	
	62.7	58.5	
Average	62.6	58.9	57

Specimen (B66) heated by induction to a surface temperature of 1500 degrees Fahr. (815 degrees Cent.) in 40.6 seconds (power input to converter 12 kilowatts); held at temperature for 30 seconds and water-quenched.

had been heated to 1525 degrees Fahr. (830 degrees Cent.) in a furnace and water-quenched. The three sets of hardness readings are compared in Table III. Before the third set of readings was taken, at least 0.01 inch was ground off the surface to insure the elimination of any decarburized metal.

Disks ⅜ inch thick were cut from one of the bars and subjected

Table IV
Hardness Values Obtained on Furnace-Heated and Quenched Disk Specimens

Time at Temperature	Quench	Hardness, Rockwell C
5 minutes	Water	55.5
5 minutes	Oil	55.5
25 minutes	Oil	54.5
60 minutes	Oil	54.5
60 minutes	Water	55.0

to various times at 1525 degrees Fahr. (830 degrees Cent.) in an electrically heated furnace and then quenched. In all cases the hardness was substantially the same, as shown in Table IV.

Frequency—The frequency at which the specimens were hardened was found to be 355,000 cycles, as measured by a wavemeter. This frequency changed less than 5000 cycles when the specimens passed through the critical temperature.

Discussion

General—A steel of high hardenability, S.A.E. 4340 (ANA Specification AN-QQ-S-756), is used in this investigation so that the depth of hardening obtained will be characteristic of the induction heating cycle used and will not be limited by the maximum depth to which the steel is capable of hardening.

An additional advantage in the use of this steel lies in its ability to fully harden in a 1-inch diameter bar when oil-quenched. The use of the oil quench greatly decreases the possibility of stresses due to quenching and makes it possible to distinguish between the effects caused by induction heating and those caused by a drastic water quench. This will be discussed in more detail under the heading "Superhardness".

A published isothermal transformation diagram (1)[1] for a steel very close in actual composition to the steels used in this investigation shows that the time for the beginning of transformation is greater than 10 seconds at temperatures above 500 degrees Fahr. (260 degrees Cent.). Consequently, no transformation should take place during the 3 seconds or less necessary to transfer the heated specimen from the coil to the quench.

This investigation is intended to show the metallurgical changes and characteristics of steel heated by induced high frequency currents. It is not intended to give specific information as to the

[1]The figures appearing in parentheses pertain to the references appended to this paper.

means of hardening parts. The numerical values are given for the purposes of illustration and comparison of the various factors involved and must not be taken as recommended values for commercial use.

Temperature Measurement—Both optical pyrometers and thermocouples have been used to measure temperature in the published work on induction hardening. Under suitable conditions, either instrument will give the same result. For the measurement of the surface temperature of induction-heated pieces, only an optical or other type of radiation pyrometer is believed to be suitable. While no appreciable effect is produced in the thermocouple wires by the high frequency field (2), the irregularity where these wires are spot welded or peened to the surface of the specimen tends to cause sufficient localized overheating to make the observed temperature readings of doubtful accuracy. The use of thermocouples spot welded to the surface of the specimen is one of the factors which makes the results reported by Martin and Wiley (2) open to question.

The results obtained when the thermocouple wires are carried on the inside of the piece during heating, as described by Konold in his discussion of the paper by Benninghoff and Osborn (3), are accurate but the temperature obtained by this method is not surface temperature.

It is, of course, true that the accuracy of results obtained with an optical pyrometer may be questionable when the temperature is rapidly increasing. Under this condition, however, a thermocouple would also be expected to give erroneous results both because of the severe localized overheating at the point of attachment of the thermocouple caused by the high power input and because of the lag inherent in most types of indicating or recording instruments.

With the slower heating rates used in the present investigation, the temperature readings obtained with an optical pyrometer are believed to be reasonably accurate. The temperatures reported are the "apparent" (uncorrected) temperature readings given by the disappearing filament type instrument used.

In the temperature range and under the conditions used in these tests, the difference between the "apparent" and the true temperature is small. The values given for the spectral emittance of iron oxide at 0.65μ range between 0.63 and 0.98. When the apparent temperature is 1500 degrees Fahr. (815 degrees Cent.), the true temperature is calculated to be 1540 degrees Fahr. (838 degrees Cent.), using the

lowest emittance value given. The difference between the apparent and true temperatures becomes less if the emittance value is actually higher than the 0.63 value assumed. A hardening temperature range of 1475 to 1550 degrees Fahr. (800 to 845 degrees Cent.) is specified by ANA Specification AN-QQ-H-201, "Heat Treatment of Steel (Aircraft), Process for," for S.A.E. 4340 steel. The specimens heated to an apparent temperature of 1500 degrees Fahr. (815 degrees Cent.) are therefore heated within the correct hardening range, even though the value of the emittance under the test conditions, and consequently the true temperature, is not exactly known.

At 1900 degrees Fahr. (1040 degrees Cent.), the correction to be added to the apparent temperature is only 70 degrees Fahr. when the emittance is as low as 0.60, a very conservative value.

Hardness Patterns—The data show that the extent of hardening in the longitudinal direction is primarily a function of the coil length only for the range of temperatures and power input investigated. In all cases, the length of the hardened surface is less than the coil length. This difference of course depends on the design and size of the coil used, but it illustrates the point that the length of the hardened zone does not necessarily coincide with the length of the inductor used.

In passing it should be remarked that the extent of the hardened zone is not indicated by the temper colors on the surface. These temper colors usually accurately portray the extent and shape of the induction coil.

During induction heating, the specimens rest on a transite board ½ inch below the bottom of the inductor coil. Since this distance is so short that normal conditions of heat flow do not exist during the entire heating cycle, only the top half of the specimen is used for the investigation. The specimen extends at least 1½ inches above the coil and the conditions in this part of the specimen are believed comparable to those existing in the commercial induction heating of certain areas or zones of a part.

It is generally recognized that the surface properties of a surface-hardened part extend for only a short distance under the surface. When, however, a piece is to be deep or through hardened, it is not always appreciated that the transition between the hardened and unaffected zones may be cone-shaped instead of a straight line extending inwardly perpendicular from the surface. At least one service failure in aircraft has been definitely attributed to failure to recognize

this fact. While the surface-hardened zone extended beyond the highly stressed area, the inner cone of unheat treated metal extended well into the stressed area and the part failed.

This cone-like transition is not eliminated under any of the test conditions used. The nearest approaches to a straight line transition perpendicular to the surface are obtained by holding the specimen at a surface temperature of 1500 degrees Fahr. (815 degrees Cent.) for 30 seconds or by raising the surface temperature to 1900 degrees

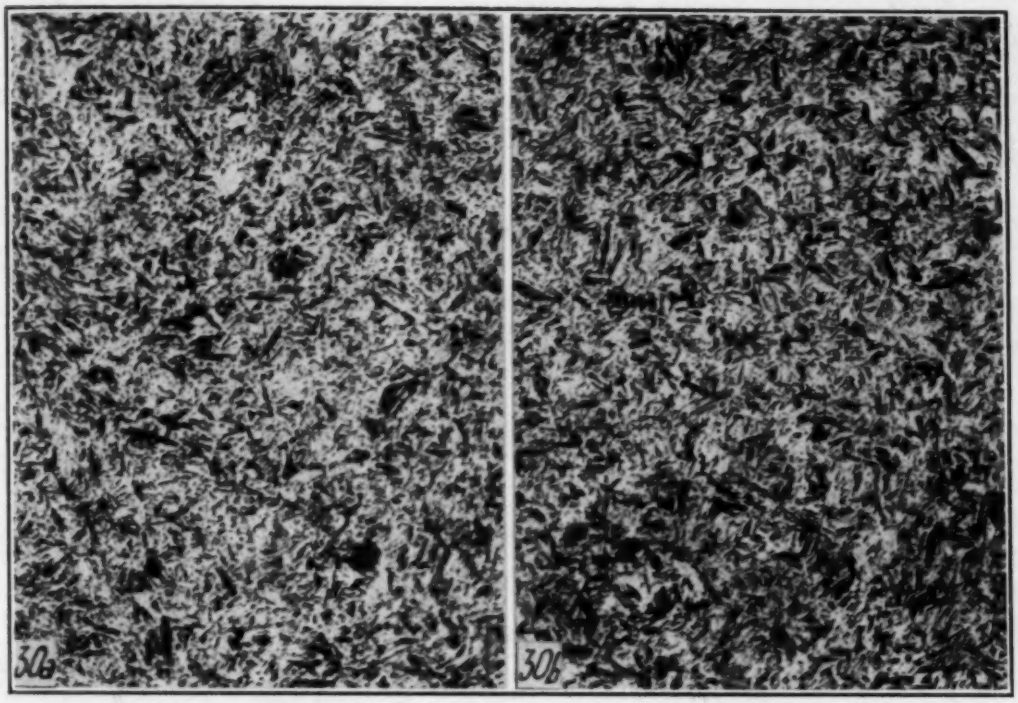

Fig. 30—Similarity of Martensitic Structures Produced by Two Methods of Heating. Induction-heated on left; furnace-heated on right. Oil quench in both cases. Nital etch. × 500.

Fahr. (1040 degrees Cent.). However, the first method would be difficult to apply commercially and the second produces serious coarsening of the grain structure. It may prove to be impossible to entirely eliminate this cone effect since the heat is produced near the surface by the induced currents and must flow into the interior. Simultaneously, heat is flowing longitudinally from this heated zone into the cooler ends of the metal outside the field of the coil. The resulting thermal equilibrium may be such that all the interior of the piece near the ends of the coil will never reach the hardening temperature.

In some cases the hardness of a narrow circumferential band at the ends of the induction-hardened area is found to be lower than

the original hardness of the steel. This effect is due to the tempering of the metal adjacent to the zone heated above the critical (hardening) temperature. As shown by the longitudinal hardness surveys, the softening varies from 0 to 20 points Rockwell C, depending on the original hardness of the material. Since a comparatively low temperature is sufficient to temper a fully hardened steel, the maximum hardness decrease and maximum width of band are found on the specimens having the highest original hardness. The band is narrower and the hardness drop less on the softer (normalized and tempered, or quenched and tempered) specimens. Naturally no drop in hardness below the original value is found in the specimens induction-hardened from the spheroidized condition.

This soft band can cause serious difficulty unless its presence is recognized and due allowance for it made. Almen (4) points out the adverse effect of this band on the fatigue strength. He also points out that a similar soft transition zone may exist between the hardened surface and the partially hardened core. With the sample size and test conditions used, this latter phenomenon is not observed in this investigation. An approach to it is noted in the softening (to Rockwell C-35) of the interior of the fully hard (Rockwell C-55 to 56) specimens which are subsequently surface-hardened by induction.

Metallographic Structure—The hardened surface zone of the induction-hardened specimens consists of the usual martensitic structure, in no way different from the martensitic structure found in furnace-heated specimens. For ready comparison the structure obtained near the surface of a sample induction-heated to a surface temperature of 1500 degrees Fahr. (815 degrees Cent.) is shown in Fig. 30, together with the structure found in a piece of the same steel 1 inch in diameter, furnace-heated to 1525 degrees Fahr. (830 degrees Cent.) and oil-quenched.

Although the similarity of the martensitic structures in induction and furnace-heated steels is denied by Osborn (5), (6), (7); Vaughn, Farlow and Meyer (8); Martin and Wiley (2); and Konold in the discussion of Benninghoff and Osborn's (3) paper, all report this similarity. Included in these investigations are plain carbon (0.17 to 1.10 per cent carbon) and WD 4150 alloy steels hardened at frequencies of 3000, 9600, or 530,000 cycles. The preponderance of published evidence seems to substantiate the finding of this investigation.

This martensitic structure coarsens with increasing temperature

in the same way as furnace-heated steels. Here again the result is in direct conflict with the statement of Benninghoff and Osborn (3) that "due to short heating cycles, induction heating is not bound by the usual limitations of coarsening temperatures and we can overcome the reticence of certain structures to go into solution by heating to temperatures several hundred degrees above the upper critical with no deleterious effects." The coarsening temperature of the spheroidized specimens is higher than that of the normalized and tempered specimens. This may be due to the inhibiting effect on grain growth produced by the carbide particles in a way similar to the effect produced by alumina inclusions.

The structure of the transition zone in the induction-hardened specimens is similar to the structure of steels furnace-heated to temperatures below or within the critical range.

These results also illustrate the importance of the steel being in a suitable condition to respond to induction hardening. The normalized and tempered specimens, in which the carbide particles are small and well distributed, harden readily when heated for a short time to a surface temperature slightly above the critical. Longer times and higher temperatures increase the depth of hardening.

The spheroidized specimens (carbide particles large) do not respond as readily to induction heating. A holding time in excess of 30 seconds at 1500 degrees Fahr. (815 degrees Cent.) or a surface temperature of 1800 degrees Fahr. (980 degrees Cent.) is necessary for full hardness to develop. Microscopic examination shows that all the carbide particles near the surface do not disappear until a surface temperature of 1900 degrees Fahr. (1040 degrees Cent.) is reached.

While the spheroidized samples tend to respond more slowly, the structures found, except for the presence of the undissolved carbide particles, are the same as those found in the normalized and drawn samples.

An excellent correlation is obtained between the microscopic structures and the transverse hardnesses. The drop in hardness, however, gives a clearer indication of the exact location of the transition zone than the changes in structure.

Superhardness—One of the controversies in connection with induction-hardened parts concerns the existence of superhardness. Tran and Osborn (5) obtained surface hardness ranging from 2 to 6 points Rockwell C higher than the values given by Burns, Moore

and Archer (9) as the maximum hardness obtainable for steels of given carbon content. The results obtained by Vaughn, Farlow and Meyer (8) confirm the existence of these higher hardness values. These latter authors, however, conjecture that this behavior is produced by internal stresses in the piece. Martin and Wiley (2) also find higher hardness on induction-hardened specimens which they attribute to high compressive stresses, less retained austenite and carbon segregation effects. In all of these cases, the test pieces are water-quenched.

With the few exceptions discussed below, the specimens used in this investigation are oil-quenched. The hardness obtained is very uniform from specimen to specimen and ranges between Rockwell C-55 and 56. For steel of 0.40 per cent carbon content, Burns, Moore and Archer (9) give a value of Rockwell C-59 to 60 as the maximum attainable hardness. This value is based on extremely rapid quench-

Table V
Power Input to Specimen

Converter Input kilowatts	Input to Specimen kilowatts per square inch
12	0.92
15	1.15
16	1.22
18	1.37
19	1.45
28½	2.18

Input to specimens calculated assuming efficiency of converter to be 60 per cent.

ing into iced brine of quite small specimens. For quenching rates obtainable in commercial practice, Boegehold (10) gives a minimum value of Rockwell C-54.5 for steel of this carbon content. The hardness values obtained on furnace-heated and oil- or water-quenched disks of the steel used in this investigation are identical with those obtained on the induction-heated and oil-quenched specimens, and all the values are within the range which corresponds to fully hardened steel of this carbon content.

On the specimens of the same steel induction-heated under the same conditions to produce a hardened surface zone but water-quenched, a Rockwell hardness of C-62 to 63 is obtained in most of the hardened layer. This value is close to the 64.2 value obtained for this type steel by Tran and Osborn (5). When the quenched specimen is cut into several parts, the hardness decreases 1½ to 2 points Rockwell C. This difference is obtained by comparing the hardness

values obtained within $\frac{1}{16}$ inch of each other on the piece before and after cutting. While the hardness of the cut specimen does not decrease to Rockwell C-56, the hardness characteristic of this steel when oil-quenched, the drop in hardness shows a definite change has taken place.

The 325 degrees Fahr. (165 degrees Cent.) to which one of the induction-heated and water-quenched specimens was heated is a high enough temperature to at least partially relieve any stresses present without softening the metal by tempering. The fact that the decrease in hardness found after this stress relief treatment is about the same as that found on cutting the specimen indicates that internal stresses are a factor in the high hardness found in the water-quenched specimens. Additional confirmation is found in the fact that the water-quenched specimens warp and grab the cutting wheel, while the oil-quenched specimens do not. Vaughn, Farlow and Meyer (8) report similar difficulty.

These data lead to the opinion that the so-called "superhardness" found in some induction-hardened steels is not due to any inherent difference in structure produced by the induced currents, but may be related to the internal stresses produced by the differential heating and quenching. The fact that the hardness values obtained are dependent upon the method of quenching, and not upon induction heating, is considered to substantiate this.

Similar opinions have been expressed by Vaughn, Farlow and Meyer (8) and recently by Osborn in his discussion of the work of Martin and Wiley (2).

Frequency and Power Input—Much discussion in the published literature is concerned with the effect of frequencies and power input per square inch of sample. For ready comparison with previously published data, the calculated power per square inch of specimen corresponding to the various power inputs to the converter is given in Table V. In making these calculations, the values of power input to the converter as supplied by the manufacturers are used and the efficiency of the converter is assumed to be 60 per cent. This figure is somewhat more conservative than the 70 per cent efficiency "under favorable conditions" given in Fox's discussion of the paper by Benninghoff and Osborn (3).

The depth of metal heated by the induced currents depends on the frequency, but the high rate of heat conductivity through the steel also plays an important part in the depth of metal hardened.

Calculations based on the Gurney-Lurie chart for a long cylinder (11) show that the metal 0.1 inch beneath the surface of a 1-inch diameter bar attains a temperature of 925 degrees Fahr. (495 degrees Cent.) in 1 second, assuming the surface instantaneously reaches a temperature of 1500 degrees Fahr. (815 degrees Cent.). Since a finite time is required for the surface to be heated to 1500 degrees Fahr. (815 degrees Cent.), the temperature differential between the surface and points in the interior will actually be less than that indicated by this calculation. Obviously as the surface temperature is increased above 1500 degrees Fahr. (815 degrees Cent.), the temperature of the interior will increase and more of the metal will be heated above the critical temperature. While the depth of heating varies inversely as the square root of the frequency, it also varies directly as the square root of the resistivity and inversely as the square root of the permeability (12). The numerical value of the permeability below the critical temperature is of the order of magnitude of hundreds, if not thousands, and decreases to unity above the critical. The resistivity also varies with temperature, but the magnitude of variation is much less. This causes the maximum heating effect to shift toward the interior as the surface metal reaches the critical. These factors, together with the heat conductivity, make it seem probable that the effect of frequency on the depth of hardening has been overemphasized.

It is undoubtedly true that the power (heat) input per unit area is important. In fact it may well be that the high heat inputs per unit area and the ability to localize and control the shape of the heated area prove to be the only characteristics distinguishing induction heating from furnace heating.

However, it seems that much of the published data on power input per unit area are calculated, rather than actually measured values, since simple calculations using the specific heat, the density, the given depth of hardening and power input show that either the heating time is only a fraction of a second or the surface is greatly overheated. In almost all cases, the heating times used are reported to be greater than 1 second.

Using the same method of calculation, a heating time of 52 seconds is found for the specimens used in this investigation through heated with the lowest power input. This calculation assumes an efficiency of 60 per cent for the converter, as previously discussed. The times actually observed for the eleven specimens range from 44.2

to 54.4 seconds with an average of 48.5 seconds. For the through hardened specimen heated with a power input to the converter of 19 kilowatts, the calculated time is 33 seconds compared with an observed time of 40 seconds. Considering the assumptions and approximations necessary in making the calculation and the uncertainty as to the exact temperature to which the specimen is heated, the agreement between the calculated and observed values is very good.

Summary

The response to heat treatment, the resulting hardness, and the metallographic structure of S.A.E. 4340 steel are the same whether the specimens are heated in a furnace or by induced high frequency electric currents. Induction heating differs from furnace heating in that the heat may be localized in any desired area and in that a much higher heat input per unit area in unit time can be obtained in many cases. The depth of hardening is controlled more by the power input and time of heating than by the frequency of the induced current.

The so-called "superhardness" found in some induction-hardened steel is apparently related to the presence of internal stresses and is not an effect produced by the induced currents.

Acknowledgments

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Appendix I
Calculation of Heating Rates

The Gurney-Lurie charts for various shaped solids (11) show the relation between the "relative time" ratio X, and the temperature difference ratio Y. Mathematically

$$X = \frac{k\theta}{\rho c_p r_m^2}$$

and $$Y = \frac{t^1 - t}{t^1 - t_b}$$

where k is thermal conductivity of the solid, B.t.u./(hr.) (sq.ft.) (deg. F. per ft.)

θ is time (hours) from start of heating

ρ is density of solid, pounds per cubic foot

c_p is specific heat of solid, B.t.u./(lb.) (deg. F.)

r_m is normal distance from midplane to surface, feet

In the case of cylinders, the radius

t^1 is the temperature of surroundings, deg. F.

t is the temperature at position n at time θ, deg. F.

t_b is the original temperature of solid, deg. F.

The value for k for steel varies from 26 at 212 degrees Fahr. (100 degrees Cent.) to 21 at 1112 degrees Fahr. (600 degrees Cent.). For the temperature range of 70 to 1500 degrees Fahr. (815 degrees Cent.), the value of k will be taken as 22.

The value for the density is taken as 489 lbs./cu.ft.

Although the value of c_p changes with temperature, a constant value of 0.216 B.t.u./lb. is used in these calculations.

Since the cylinder is 1 inch in diameter, r is $\frac{0.5}{12}$ ft.

$$\therefore X = \frac{22 \times \frac{\theta'}{60 \times 60}}{489. \times 0.216 \times \left(\frac{0.5}{12}\right)^2} = 0.033\ \theta'$$ where θ' is time in seconds.

In using the diagram, it is necessary to select a value for m, the resistance ratio. For the purposes of this calculation, it is assumed that the surface of the cylinder reaches the desired temperature instantaneously, giving the value $m = 0$.

In calculating the temperature at different distances from the surface, t_b is taken as 70 degrees Fahr. and t^1 as the desired surface temperature.

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DISCUSSION

Written Discussion: By J. R. Konold, metallurgist, Armstrong Cork Company, Lancaster, Pa.

The author is to be commended on presenting some excellent data on the heating ratio and metallurgical characteristics of steel heated by 355,000 cycles.

While the writer has had no experience with induction heating of steel other than 9600 cycles, we have encountered some interesting information in heating a modified NE 8665 steel and we wonder if the author could predict whether the same conditions would exist if 355,000 cycles were used to heat this steel.

We had three heats of steel of the following chemistry:

No.	C	Mn	P	S	Si	Ni	Cr	Mo
1	0.64	1.22	0.015	0.024	0.29	0.55	0.58	0.13
2	0.63	0.94	0.018	0.022	0.31	0.55	0.45	0.12
3	0.65	0.81	0.017	0.020	0.29	0.58	0.56	0.13

The Jominy Hardenability is as follows, Rockwell "C" at 16th:

	1	2	4	6	8	12	16	20	24	28	32
1	65	65	65	64.5	64.5	63.5	62.5	62	60	58.5	56
2	62	62	62	62	61.5	59	54	48	44	42.5	40
3	62	61.5	60.5	59	57.5	54	51	48	45	43	42

The samples were ¾ inch round. We were aiming for a 5 to 7 Shepherd grain size fracture number and full hardness for both center and surface.

Samples for each heat were given a lamellar and a spheroidized anneal similar to Figs. A and B.

A time-temperature curve was made with No. 30 gage chromel-alumel wire inserted in the center of the sample and connected to a magnetic oscillograph.

The power input, holding period between heating and quenching and quench-

ing time were held constant for all samples but the time of heating was varied with the following results. All samples were water-quenched.

Heat No.	Heat Time Seconds	Temperature Degrees Fahr.	Rockwell "C" Hardness Surface	Center	Fracture Number
1 Spheroidized	90	1690	64/64.5	63/64.5	4–6
	85	1660	64/64.5	63/64	4.5–6
	*80	1640	63/64.5	63/64.5	5–6.5
	75	1590	63	63/64	5.5–6.5
	70	1560	60/62	61/62	6.5–7
	60	1425	58/61	60/61	7.5–8
1 Lamellar	90	1690	64/65	63/65	4–5.5
	85	1660	64/65	63/65	4.5–6
	*80	1640	64/65	63/64.5	5–6.5
	75	1590	63/65	63/64	5–6.5
	70	1560	62/64	62/63	6–6
	60	1425	59/64	62	6–6.5
2 Spheroidized	90	1690	63.5/65	63/65	3.5–6
	85	1660	64/65	63/65	4–6
	80	1640	64/65.5	63/65	4.5–6
	*75	1590	64/65.5	63/65	5.5–6.5
	70	1560	62/65	63/64	6–7
	60	1425	61/64	62/63	7.5–8
2 Lamellar	90	1690	63/65	64/65	4–6
	85	1660	63/65	64/65	4–6
	80	1640	64/65	64/65	4.5–6
	*75	1590	64/65	63/65	5–6.5
	70	1560	63/65	63/64	5.5–6.5
	60	1425	63/65	63/64	6–7
3 Spheroidized	90	1690	64/65	64/65	4.5–6
	*85	1660	64/65	64/65	5–6
	80	1640	63.5/65	64/65	5–7
	75	1590	64/65	64/65	5.5–7
	70	1560	64/65	63/65	6–7.5
	60	1425	63/65	63/65	7.5–9
3 Lamellar	90	1690	63/65	64/65	4.5–6
	*85	1660	63/65	64/65	5–6.5
	80	1640	64/65	64/65	5–6.5
	75	1590	64/65	63/65	5.5–6.5
	70	1560	64/65	63/64	6–7
	60	1425	64/65	63/64	6–7

*Indicates temperature used for production.

Since we were interested in hardness and grain size fracture, we did not examine all specimens for microstructure. However, the ones we did examine showed a coarse martensitic structure similar to Figs. C and D.

The above data indicate the following:

1. The lamellar structure goes into solution more readily than the spheroidized as is indicated by the hardness at the lower temperature.

2. Jominy hardenability test is not always an indication of the hardness obtained by induction heating as is shown by the fact that higher hardness was secured at 1425 with heat 3 than with heat 1 or 2.

3. Chemistry plays a definite part with induction heating as is shown by the fact that the only major difference in the three heats of steel is the variation in manganese. The lower manganese gave better hardness at lower temperature than did the higher manganese.

4. We were able to use the same cycle and temperature for both the lamellar and spheroidized structure of this steel, while with 4150 we had to change our cycle for lamellar and spheroidized.

In view of the above, can the author throw any light on the following:

1. Why should the hardness temperature factor vary inversely with the Jominy hardenability curve when using induction heat, or is this a function of chemistry?

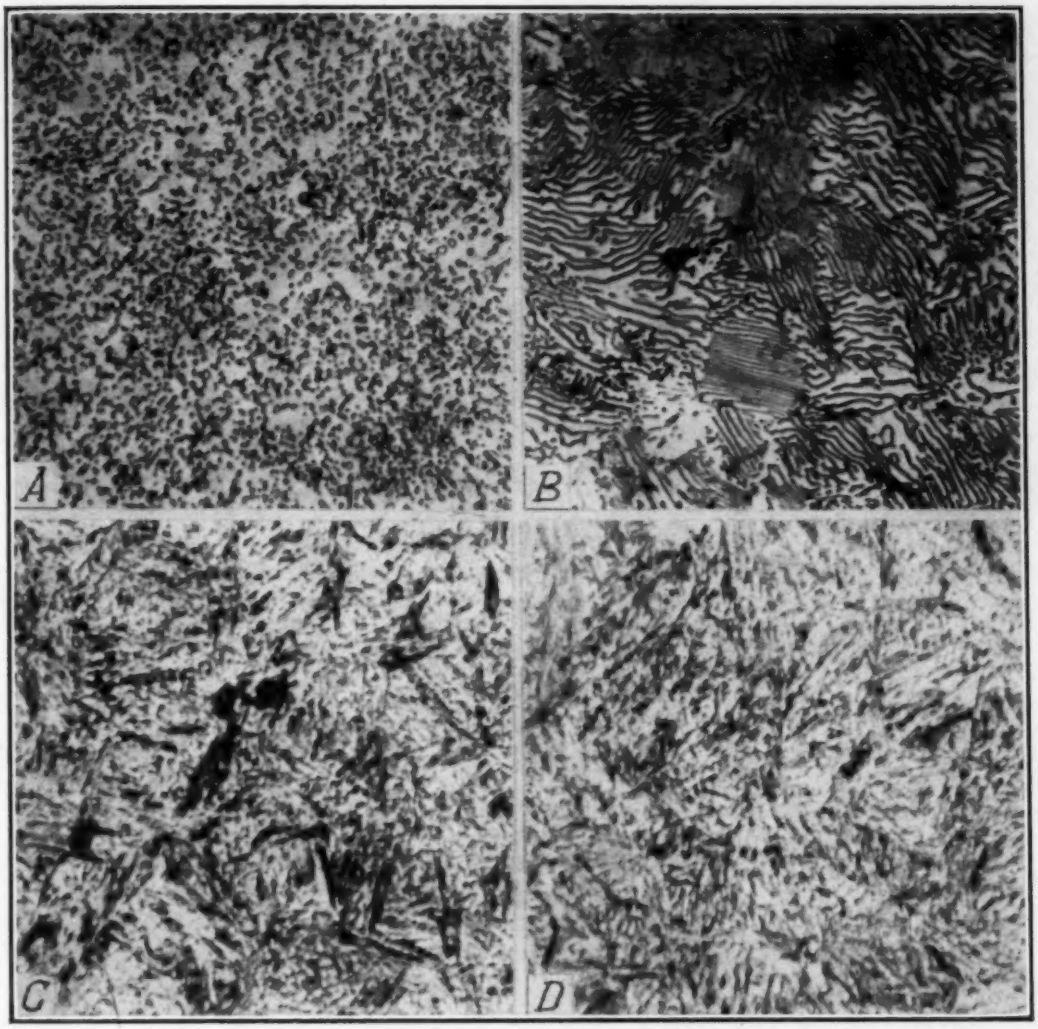

Fig. A—Heat No. 1—Spheroidized Structure. Picral etch. × 750.
Fig. B—Heat No. 1—Lamellar Structure. Picral etch. × 750.
Fig. C—Microstructure of Heat No. 1 Spheroidized Annealed. Quenched from 1640 degrees Fahr. and tempered at 300 degrees Fahr. Picral etch. × 750.
Fig. D—Microstructure of Heat No. 1 Lamellar Annealed. Quenched from 1640 degrees Fahr. and tempered at 300 degrees Fahr. Picral etch. × 750.

2. Why should one type of steel give the same hardness and grain size for both lamellar and spheroidized structure for a given cycle, and a different steel require a different cycle for each structure?

Written Discussion: By James W. Taylor, staff engineer, and Arthur E. Focke, research metallurgist, Diamond Chain and Manufacturing Co., Indianapolis.

The data contained in this paper are very interesting to us even though the type of steel used is not one which we attempt to induction harden.

However, we wonder why the samples were not quenched through the coil rather than removed and transferred to the quenching medium. It would seem that a delay of as much as 3 seconds during this transfer would be a relatively long period of time as compared with the actual time the power is on, particularly for the higher power inputs and lower surface temperatures.

Then, we found it difficult to understand why the normalized and tempered sample hardened practically completely throughout the cross section with a 12-kilowatt power setting and 1500 degree Fahr. surface temperature while the "hardened" or "hardened and drawn" samples under the same test conditions were not hardened as deeply. To aid us in our analysis of these test results, we replotted the center hardness data from the author's Figs. 2, 5, 8, 9 and 10 as shown in Fig. E.

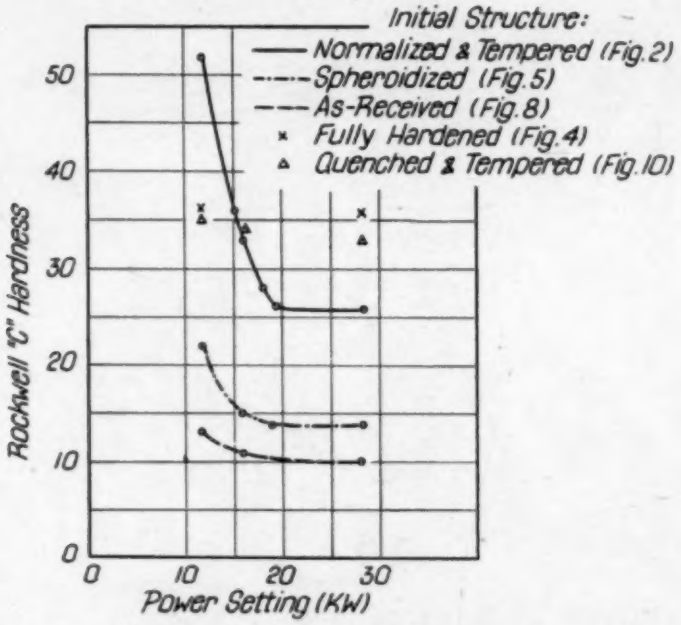

Fig. E—Center Hardness Versus Power Input at 1500 Degrees Fahr. Surface Temperature.

From this Fig. E it would seem that on the basis of center hardness the "hardened and drawn" samples gave results more nearly similar to those obtained with the "hardened" sample than with those of the "normalized and tempered" bars as reported by the author. In this connection, it would be interesting to know what tempering temperature was applied to these "hardened and drawn" samples.

Finally we question whether the author had sufficient data to justify his conclusion that "The depth of hardening is controlled more by power input and time of heating than by the frequency of the induced current."

All the tests reported in this paper were run on a spark gap converter at a frequency of 355,000 cycles. While his theoretical analysis of the effects of frequency, permeability and resistivity may support his conclusion, it would appear more cogent if it had been supported by actual comparative results using other equipment with different frequencies. It has been our experience, based on comparison between results obtained on our 60-kilowatt, 9600-cycle motor generator unit and our spark gap converter, that the frequency difference between these units determines, in a large measure, which unit should be used for specific jobs.

We hope these comments will show the author that his contribution to our knowledge of the metallurgical features of induction hardening is sincerely appreciated and that they may encourage him to use the excellent testing facilities

available at Wright Field to make further investigation in this field.

Written Discussion: By J. P. Moran, research engineer, Induction Heating Corp., New York.

The author is to be complimented on a very thorough treatment of a subject which is becoming of increasing interest to more people every day. The results obtained and reported in this paper will no doubt aid considerably in the use of induction heating as a heat treating tool.

In going over this paper, however, there are several points which we feel need clarification.

The first of these involves the effect of power input on the depth of hardening. In the use of the curves reported, it should be borne in mind that while different values of power input were used, this cannot be construed as representing values of power input into the work, but rather are the input values of power fed to the converter from the power lines. It is true we can assume for a given power input to the converter, everything else remaining constant, that the power input to the work would be some value less than the input by an amount equal to the losses in the circuit. In using several values of input to the converter and plotting results of hardening, the inference is made that they are representative of power input to the work proportionate in every case to the power input to the converter. This is not true, since the losses do not remain constant, the circulating current is not constant and therefore the power available for the work cannot be a function of power input alone. It would have been better to base the results on circulating current in the work coil.

We also believe that the curves shown in Fig. 4 do not indicate the effect of power input on depth of hardening. They do, however, indicate that the maximum power available was insufficient to provide 1600 degree Fahr. (870 degree Cent.) surface temperature in a sufficiently short time to prevent conduction of heat and subsequent hardening of the entire cross section of the specimen.

Before passing from this subject there is one other point and this has to do with the time required to heat to various surface temperatures. If we remember that the radiation loss is proportional to the fourth power of the temperature difference, it is inconceivable that, with constant power input, the time to reach the temperatures given would be a linear function of those temperatures. We believe the results reported are due to the method of measuring temperature at one spot and that while that spot assumed the required temperature in the time reported, the band being heated became smaller for each increase in temperature.

Summarizing all the foregoing brings us to the following conclusion: We agree with the author that the power input to the work, not to the converter, has a definite effect on the depth of penetration, provided only that there is sufficient power available to overcome the loss by radiation, and to realize the required surface temperature in a period of time short enough to prevent any appreciable conduction of heat into the core.

In the section of this paper on "contour of hardness pattern", we believe the results are not indicative of good induction heating practice but rather are those which are usually obtained with multi-turn coils of uniform spacing. It should be pointed out that the effects produced herein are typical of multi-turn coils with wide spacings between turns, but that such results are not necessarily obtained when optimum coil design is used. The author describes the coil used in

these tests as being "5¾ complete turns with a ⅛-inch spacing between turns" and also that "the temperature readings were made with a disappearing filament-type optical pyrometer focused on the specimen between the center turns of the coil". With the coil design used, the magnetic field will have its greatest concentration at the center vertically with the field pattern, tapering off in intensity toward both ends. If, as in these tests, the temperature is read at the center of the coil, this will represent the temperature existing at the "hottest" part of the coil and will not indicate the average condition of the sample. If the sample is then quenched out, those parts of the specimen included within the end turns of the coil, not having received as much energy as the center, have less heat to dissipate into the core and the resultant cone pattern obtains. This is obvious from the series of pictures in Fig. 14. It can be seen from these that the cone pattern is more pronounced at 1500 degree Fahr. (815 degree Cent.) surface temperature than it is at 1900 degrees Fahr. (1040 degrees Cent.). In obtaining 1900 degree Fahr. (1040 degree Cent.) surface temperature as measured at the center of the specimen, considerably longer time is required than that to obtain 1500 degrees Fahr. As a result of this, there is more time for the less efficient end turns of the coil to raise the temperature of that part of the specimen enclosed by them. The hardness pattern thus more nearly approaches a straight line than the 1500 degree Fahr. (815 degree Cent.) sample in which there was less time for the end turns to become effective.

We have found that in through heating of bar stock using induction heating equipment operating at 375,000 cycles, it is possible to obtain a pattern which ends almost in a straight line level with the top of the coil if certain design principles in coil making are followed. These can be of two general types. The first and the most difficult to make involves winding the coil with uniform spacing but with a varying inside diameter which is greatest at the center and least at both ends. It will be readily seen that the difficulties in fabricating this type of coil render it impractical. The second method, involving merely variation in the spacing of the turns, is much easier to accomplish. More turns are concentrated at the ends of the coil with very close spacing between them. While there are no practical formulae which can be used to indicate the spacing between the turns, fortunately the effect is not too critical and can usually be accomplished in one or two tries.

On the matter of the effect of frequency on depth of hardening, without attempting to go into all the various factors involved in the author's discussion of this subject, we think it should be pertinent that since, regardless of the frequency used, the other factors (namely, heat flow, change in resistivity, change in permeability, etc.) are always present and the established fact that the depth of current penetration (not the depth of heating as given by the author) is inversely proportional to the square root of the frequency at these frequencies, it follows that the shallower the initial path of the current, the shallower will be the final heated zone for the same power input to the work at a short heating time. The other factors are functions of the temperature and therefore cannot be considered as having much bearing on the effect of frequency on depth of heating.

In his discussion of the effect of power input per unit area the author points out that the usual calculations, using the specific heat, the density, the given depth of hardening and power input, show that either the surface is badly over-

heated or the heating time is only a fraction of a second, whereas, in all cases, heating times reported are always greater than 1 second. We believe that the calculation using the above factors represents only theoretical conditions and does not include the effects of radiation and conduction losses at the temperatures involved and also that these two factors require considerably more power than that theoretically necessary to bring the given mass of material to the required temperature. It is important to point out that the author's observation of heating times against those calculated shows very close agreement, despite the fact that apparently from the pictures shown in the section on the contour of the hardened pattern, only a small percentage of the total mass included within the coil was heated to the required temperature.

In conclusion, we agree with the author's statement that "the depth of hardening is controlled more by power input in time of heating than by frequency of the induced current" if this is expanded to include "at frequencies of the same order" and if power input is interpreted as power into the work. We disagree that the longitudinal effects recorded here are always necessarily to be found in induction heating.

Written Discussion: By H. B. Osborn, Jr., research director, Tocco Division, Ohio Crankshaft Co., Cleveland.

While the author is to be commended for the painstaking effort in developing the data in this paper, we feel that most of it is highly inconclusive due to the limitations imposed upon the studies by the type of induction heating equipment used. We fail to see how a thorough understanding of induction heating can be achieved from experiments involving surface hardening treatment wherein power inputs per unit of surface are considerably less than those used in practice, and specimens treated rather small in size, thus producing samples not truly representative of the induction heating process. Furthermore, the use of a multi-turn coil of uncertain dimensions can hardly be considered the proper mechanism with which to conduct such tests. The magnetic flux produced and the irregularity of quench caused by the turns are objectionable. We note that the author removed the samples from the inductor for manual "dunk" quenching and while this eliminates the objection just raised to quenching through a multi-turn coil, it introduces errors in quenching temperature despite care exercised in checking the time interval involved in moving the part. No study of induction heating should be considered without the use of an accurately machined inductor block of fixed dimensions to produce a known flux pattern and reliable quench coming through the inside face.

In order to present our points for ready reference, we will tabulate in accordance with subdivisions of Mr. Poynter's paper:

Abstract—The transition between the hardened zone and the unaffected metal may have been cone-shaped in the studies made by Mr. Poynter, but this is due to the fact that he used heating cycles which are not representative of production induction surface hardening applications. A multi-turn coil with turns spaced wider in the center or with greater coupling would have minimized the cone-shaped result although, as pointed out above, a multi-turn inductor is not dependable. A solid, single-turn inductor, with a portion of the central part of the internal face accurately recessed to provide less coupling, will flatten out the magnetic field and produce a flat hardness pattern, not cone-shaped, but with

only a slight radius at each end where it comes to the surface.

We quite agree that the effect of frequency on depth of hardness has been overemphasized and that heat flow by conduction must be given careful consideration. We refer specifically to data presented by the writer in other publications and currently in his paper presented during the lecture series at the 1945-46 A.S.M. convention.

We have also, in previous publications, indicated that the superhardness generally associated with induction hardening is related to internal stresses, but data by the author of this paper have not explained the total differential of hardness on this basis.

Experimental Procedure—The inductor used in this experiment comprised 5.75 turns to 0.25-inch copper tubing spaced 0.125 inch apart. Such an inductor will not produce a uniform field since the area beneath the middle turns will be subjected to a much greater intensity of magnetic flux. This accounts, therefore, for the cone-shaped pattern, particularly since the heating times were so excessive.

Manual removal of the specimen from the inductor coil for quenching is not in concurrence with recognized production methods for surface hardening wherein the part is automatically quenched in position.

Power data are highly unreliable since the 60-cycle input in this case does not have a direct relationship to the high frequency input to the work. We cannot assume a uniform input of energy to the piece during the heating cycle. Due to the change in electrical characteristics of the material during an increase in temperature, it does not absorb the same amount of energy during the entire heating cycle, even though the 60-cycle power input to the oscillator may remain the same. For a given power input to the work at the beginning of the heating cycle, we find a decrease with temperature rise and the balance of energy is absorbed as heat in the spark gaps, condensers, leads, etc. Furthermore, radiation losses during long heating cycles as used in the experiments are not reliably accounted for.

The drawing or softening effect noticed beneath the induction-hardened surface of a bar previously heat treated is generally quite well recognized, but one of the many advantages inherent with the use of induction hardening is that it permits the use of ordinary carbon steels in a normalized condition which will not show this effect. While the drawing effect is related to the type of steel and core hardness and structure, it is not encountered with materials at hardnesses below approximately Rockwell C-28 to C-30.

Longitudinal Extent of Hardening—No perceptible difference may have been noted in the extent of the hardened zone for the normalized and tempered specimens for various inputs due to the long heating cycles, but with higher power inputs per unit of surface area, such a condition would not exist. Furthermore, a noticeable difference in pattern would have appeared under fixed conditions using short cycles with spheroidized samples as compared to pearlitic structures, due to the difference in the specific resistance of the material.

Contour of Hardened Metals—The lack of difference in results in this case is due entirely to the fact that power input per unit of surface area is below general minimums and results in long heating times which permit a large amount of heat flow by conduction, not only radially, but longitudinally.

Metallographic Examination—Here again, the long heating times used have resulted in erroneous conclusions.

Hardness—Hardness values, we agree, depend upon the method of quenching, but in all production induction hardening work where the quench water is sprayed through the inside face of the inductor, we find no difficulty in cooling the inductively heated annular zone at a rate considerably in excess of the critical cooling rate, thereby satisfying the metallurgical requirements for quenching. As far as we are concerned, quenching in this manner is considerably better and gives hardness values indicated by the writer in contrast to those generally expected because of limitations imposed by quenching rates obtainable in commercial practice. The author admits that the hardness of the specimen cut to relieve the internal stresses does not decrease to the Rockwell C value obtained by conventional methods of hardening. However, he offers no explanation for this remaining increment, small as it may be.

Hardness Patterns—Mr. Poynter states that in all cases the length of the hardened surface is less than the coil length. While this may be true of his observations, we should like to point out that with the particular type of inductor and coupling, and with high frequency, it is possible, under certain conditions, to produce an inductively heated zone wider than the inductor itself. Furthermore, with long heating rates permitting heat flow by conduction longitudinally, we can obtain similar results accompanying, obviously, a deeply heated pattern.

We believe that some qualification of the expression "deep hardening" should be made. Insofar as we are concerned, deep hardening is hardening which produces a depth generally twice the accepted standards for the frequency employed. That is to say, using 10,000 cycles we anticipate a hardened depth of 0.100 inch for short heating cycles and high power densities per unit of surface area. This depth can, of course, be decreased by using even higher power densities and shorter heating times, or increased by using lower power inputs and longer heating times. With 10,000 cycles and under conditions just stated in the foregoing sentence, we consider deep hardening when the depth extends to approximately 0.25 inch. Please note that in general we do not attempt in excess of 0.25 inch on most steels induction-hardened since the hardenability limitations usually make it impractical to exceed this value. Mr. Poynter states that the nearest approach to a straight line transition perpendicular to the surface is attained by a 30-second heat at 1500 degree Fahr. (815 degree Cent.) surface temperature or by a 1900 degree Fahr. (1040 degree Cent.) heat at the surface. We see no particular difficulty in developing the first method commercially and we certainly take exception to the statement that the second method produces serious coarsening of the grain structure. Grain coarsening is admittedly dependent upon temperature, but it is also a time effect. Here again, the limitations imposed upon the experimental work of the author by the equipment preclude the possibility of reaching such temperatures without using long heating cycles. We repeat, as we have in many published articles, that induction hardening permits the use of higher temperatures without the deleterious grain growth effects. Obviously, high power inputs per unit of surface area are required to permit short heating times.

Metallurgical Structure—The similarity of martensitic structure in induction and furnace-treated steels has been denied by the writer but substantiated by others. Exceptions to this conclusion made by the author of this paper and

others to whose papers he has referred have, in every case, been the result of a study of structures produced with long heating cycles not indicative of induction heating practice.

Mr. Poynter states that the martensitic structure coarsens with increasing temperature in the same way as furnace-heated steels and is in direct conflict with statements made by the writer in previous publications. We repeat that the author of this paper has completely ignored the time factor. We can demonstrate that, for example, in the hardening of S.A.E. 1045 steel temperatures of 1800 to 1900 degrees Fahr. (980 to 1040 degrees Cent.) have been used for surface hardening with no indication of coarsening in the microstructure. This is entirely due to the fact that a short heating cycle was used totaling approximately 2.5 seconds. The first 2 seconds were sufficient to bring the steel up to usual hardening temperature of 1600 degrees Fahr. (870 degrees Cent.), but due to the coarse pearlite present from the previous normalizing operation, satisfactory carbide solution was not obtained. The additional 0.5 second producing the higher temperature was sufficient to overcome this deficiency.

Frequency and Power Input—The calculated input to specimen in terms of kilowatts per square inch is based entirely on an assumed efficiency of 60 per cent. As explained earlier in our discussion, the power input to the inductor does not remain a constant value during the heating cycle, due to the changes in the external inductance of the circuit caused by the change in electrical characteristics of the material as its temperature increases. This, therefore, will cause the calculations of the author to be in error. Furthermore, while he may assume an efficiency of 60 per cent based on 60-cycle input to a high frequency output, we should like to call his attention to the efficiency evolved in the transfer of the high frequency energy in the inductor to B.t.u.'s developed in the steel. Power input measurements should be in terms of kilowatts per square inch of surface area of inductor bore projected to the surface area of the part, but for comparisons of energy developed we will find a probable coupling efficiency of around 50 per cent, which means that it would be quite impractical to introduce more than approximately one-third of the 60-cycle input energy to B.t.u.'s developed in the work.

The depth of metal heated by induced currents depends on the frequency as presented in data by the writer on numerous occasions. However, heat flow by conduction as affected by power input per unit of surface area which controls heating time is far more important. For reference purposes, we wish to present the following three equations which will serve as a guide in the estimation of the depth of hardened area to be expected: (Temperature 1600 degrees Fahr.)

[1] $D_1 = \sqrt{\frac{4}{F}}$

D_1 = Depth of area heated by the high frequency energy—inches.
F = Frequency in cycles per second.

[2] $D_2 = \sqrt{.0015\ T}$

D_2 = Depth of metal heated by conduction—inches.
T = Heating time in seconds.

[3] $D = D_1 + D_2$

D = Depth of hardened zone—inches.

We agree whole-heartedly with the author's statement that the effect of frequency on the depth of hardening has been overemphasized. A few calculations with the above equations will show that the use of ultra-high frequencies, unless with extremely short heating cycles, is of very little advantage in producing shallow case depths. Furthermore, power requirements needed in such cases require such expensive equipment that the economics of the application is questionable.

All of the data published by the writer on power input are based on actual measured values of energy into the inductor as read by a kilowatt meter in the case of motor generator equipment, or accurately calculated from meter readings with vacuum tube oscillators. Unfortunately, it is not possible to make any reliable calculations or take any meter readings of energy input to an inductor with a spark gap oscillator.

Author's Reply

In reply to Dr. Osborn's criticisms, the purpose of this investigation is "to show the metallurgical changes and characteristics of steel heated by induced high frequency currents. It is not intended to give specific information as to the means of hardening parts" (pages 186-187 of the paper). In the absence of any accepted standards of heating time for induction heating or hardening, a range of heating times was selected so that a good degree of accuracy would be obtained in the temperature measurement and control. The advertising pamphlet, "Results with Tocco", dated June 1945, and distributed at the Ohio Crankshaft Company's display at this present National Metal Exposition, shows a tractor drive gear requiring 90 seconds to induction heat the teeth for hardening, and a tractor roller assembly requiring 2 minutes 15 seconds heating time for surface hardening. In addition Benninghoff and Osborn (page 323, loc. cit.) describe a 40-millimeter shot (WD 4150 steel) requiring 70 seconds' heating time. It therefore seems quite arbitrary to characterize the results presented in this paper as "highly inconclusive" because of the long heating times used.

Concerning the similarity of the martensitic structures produced by furnace and by induction heating and the coarsening of this structure with temperature, Vaughn, Farlow and Meyer (loc. cit.) used heating times of 2.5 to 5.5 seconds, Konold (loc. cit.) 60 to 120 seconds, Martin and Wiley (loc. cit.) apparently 7 to 15 seconds, and the present investigator, 20 to 65 seconds. These data cover most of the range of heating times generally used and would seem to completely deny the validity of Osborn's contention that the results reported in this paper and the confirmations found in the work of these other investigators are not correct because they are the "result of a study of structures produced with long heating cycles not indicative of induction heating practice" and because "the author of this paper has completely ignored the time factor". In passing, it may be noted that a photomicrograph in the paper of Benninghoff and Osborn, previously cited (Fig. 5, page 321), is captioned "Overheating has resulted in coarsening" although no heating time or temperature is given.

Dr. Osborn has also questioned the results on the ground that the heating along the length of the specimens was not uniform. No differences in the heating rate or in the uniformity of heating from end to end of the heated zone were

detected visually. Moreover, hardness readings taken every quarter inch along the specimen were quite consistent, indicating that the metal had been heated above the critical before quenching. All microscopic examinations and hardness transverse readings were made on the center of the bar where the temperature is accurately known.

Equation [1] for estimating the depth of hardened area was previously presented by Benninghoff and Osborn (loc. cit.). As pointed out in that paper "Unfortunately, however, this factor of depth has no physical significance, since it assumes that time is zero."

Naturally the cone-shaped contour at the end of a through hardened zone is undesirable and its presence may indicate poor practice. However, attention is called to this type of transition since, as pointed out in the paper, failure to recognize the possibility of its occurrence has resulted in failure of aircraft parts. That this cone-like transition can be minimized is shown by Fig. 14 and by Mr. Moran and Dr. Osborn's comments. To avoid any possible misunderstanding, the statement made concerning this effect will be revised to read "the transition between the hardened and unaffected zones may be cone-shaped instead of a straight line". Mr. Moran's discussion of the means of preventing this will unquestionably be of value to those concerned with coil design.

Concerning Mr. Moran's comments on the comparison of actual heating times vs. calculated ones, using the specific heat, density, etc., additional work has shown that the relatively close agreement between the calculated and observed heating times reported for the specimens through heated with the lowest power input is most likely fortuitous. Since the value of the electrical efficiency obtained is foreign to the primary purpose of this investigation, this aspect is not being pursued further.

The values for the power used in this paper are given as power input to the converter. The tacit assumption that the power input to the specimen is proportional to this is based on the information given in Mr. Fox's discussion of Benninghoff and Osborn's paper (loc. cit.). In this discussion it is pointed out that in spark gap converters of the design used in this investigation, the efficiency depends on the type of material being heat treated, type of coil, coupling of the work to the coil, etc. These factors are kept constant during the tests.

It is true, as Mr. Moran points out, that radiation from a body is proportional to the fourth power of the temperature difference. Although a detailed mathematical analysis of the problem is outside the scope of this paper, an approximation may be made using the equation for heat loss by radiation for a completely enclosed body, small compared with the enclosing body

$$q = 0.172\ A \left(\frac{T_1}{100} \right)^4 - \left(\frac{T_2}{100} \right)^4 e$$

where q is net rate of heat radiation, B.t.u. per hour

A is area of the surface, square foot

T_1 is temperature of the hotter surface, degrees Fahr. absolute

T_2 is temperature of the cooler surface, degrees Fahr. absolute

e is the emissivity of the enclosed body (taken as 0.8 in this case)

It is found that the loss by radiation to the surroundings is only 2.35 B.t.u. per minute per square inch when the surface is at 1500 degrees Fahr. (815 degrees

Cent.), and 4.43 B.t.u. per minute per square inch at 1900 degrees Fahr. (1040 degrees Cent.). This corresponds to a loss of 41 watts at 1500 degrees Fahr. (815 degrees Cent.), and 78 watts at 1900 degrees Fahr. (1040 degrees Cent.). Since the input to the specimen is better than 900 watts per square inch, the *difference* in the radiation losses at these temperatures is negligible.

Confirmation of the straight-line relationship between surface temperature and time (with a fixed power input) is found in the data presented in Mr. Konold's discussion. These heating time-temperature points fall along a straight line if the lowest temperature (1425 degrees Fahr.) is omitted. The failure of this point to fall on the line is readily understood when it is remembered that the rate of induction heating changes greatly at the critical temperature due to the change in the permeability at that temperature. 1425 degrees Fahr. (775 degrees Cent.) should be rather close to the critical for NE8665 steel.

In reply to Dr. Focke and Mr. Taylor, the use of oil as a quenching medium and the shielding effect of the inductor coil were reasons for not quenching through the coil. As pointed out in the paper, the delay in quenching is not long enough for any appreciable transformation to take place before quenching. The only effect produced by the delay would be to slightly increase the depth to which the heat penetrated. Incidentally, it may be pointed out that 3 seconds is the maximum delay and many of the samples were quenched in 1 to 2 seconds.

The use of the center hardness value as an index to compare the response of variously heat treated specimens to induction heating is not considered valid, since different metallurgical changes are involved. On the samples fully hardened before induction heating, the center hardness measurement represents a reduction from the original hardness due to the tempering, while, on the normalized and drawn specimens, the center hardness represents a partial hardening for the lower power inputs or longer holding times and a slight tempering for the higher power input. The quenched and tempered specimens were oil-quenched in 1.125-inch diameter from 1525 degrees Fahr. (830 degrees Cent.) and tempered for 1 hour at 1050 degrees Fahr. (565 degrees Cent.).

It would have added greatly to the value of these data if it had been possible to include actual results obtained on the specimens of the same steel induction-heated by 9600-cycle current and 500,000-cycle current. Unfortunately, the converters necessary to make this comparison are not available.

In reply to Mr. Konold's questions, it is believed that distribution, size, and stability of the carbides are factors controlling the speed and amount of hardening on induction heating. The Jominy hardenability, deoxidation practice, and chemistry of the steel affect the response to induction hardening only as they affect the carbide particles. This mechanism may be the explanation of the difference in response of lamellar and spheroidized structures in one type steel which was not noted in another type. Mr. Konold's data, obtained at 9600 cycles, are a valued addition to the data presented.

In conclusion, it is desired to thank Messrs. Focke, Konold, Moran, Osborn and Taylor for participating in the discussion. Their comments will aid in evaluating the data and conclusions presented and in indicating in what fields future research would be most valuable.

INDUCTION HARDENING AND AUSTENITIZING CHARACTERISTICS OF SEVERAL MEDIUM CARBON STEELS

By D. L. Martin and W. G. Van Note

Abstract

The present paper is an account of a series of experiments designed to help establish a better understanding of the metallurgical factors which are of importance in induction hardening of steel.

The hardenability and austenitizing characteristics of the following medium carbon steels were studied: S.A.E. 1050, 1350, 2350, 4160, and NE 9255, and the results discussed in their relation to the induction hardening characteristics.

The effect of alloying elements and microstructure on the induction hardening characteristics is described. The steels with low Ac_1-Ac_3 temperatures, little or no free ferrite, and medium to deep hardenability are ideal for surface induction hardening. Of the five steels investigated, the S.A.E. 1350 and 2350 come closest to these specifications.

Attention has been given to the effect of retained austenite and internal stresses on the hardness of the hardened steel.

Introduction

INDUCTION hardening is well established as a metallurgical industrial process. The method involves heating the steel sample by induced alternating currents to temperatures at which the rate of formation of austenite is very rapid, and then quenching to transform the austenite to martensite.

The ease of hardening a steel by induction methods will depend upon the electrical characteristics of the induction oscillator (i. e., the frequency and power output) as well as the metallurgical characteristics of the steel. Increase in the frequency or the amperage of the induced current will increase the rate of heating; and accordingly, tend to decrease the effective depth of the heated zone. For this

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reason, shallow surface hardening may require a balancing of frequency and power to obtain the desired results. On the other hand, deep hardening is generally performed more satisfactorily with low frequency generators of high power rating. It is fortunate that these generalizations are not absolute; thus, we find that there is a wide field of application for each type of induction heating equipment—be it vacuum tube, spark gap, or motor generator set. While the electrical characteristics of the induction oscillator are of importance in obtaining satisfactory induction hardening results; they are outside the scope of this paper, the purpose of which is to discuss the second group of variables which affect the ease of induction hardening a sample—namely, the metallurgical characteristics of the steel.

The hardening cycle involves two necessary operations: First, heating to form austenite; followed by quenching to form martensite. Therefore, any adequate study of the induction hardening characteristics of steels must involve a discussion of the austenitizing and hardenability characteristics of the materials. The main object of this present work was to determine the effect of alloying elements on the induction hardening characteristics of several medium carbon steels, and to compare these properties with the austenitization and hardenability of the steels.

Formation of Austenite

Mechanism—It has been established that the formation of austentite from ferrite: cementite aggregates is by a process of nucleation and growth (1-4).[1] When first formed the austenite is not homogeneous, but is composed of carbon concentration gradients which disappear only after relatively long holding periods at the heat treating temperatures commonly used.

Effect of Temperature—The time to transform completely aggregates of ferrite and cementite into austenite will depend upon the reaction temperature, composition of the steel, and the initial microstructure. Quantitative data on the effects of these factors are lacking except for some very useful information in a paper by Roberts and Mehl in which they present their observations on the rate of formation of austenite in eutectoid steel under varying conditions of microstructure and temperature (1). Their time-temperature isotransformation diagram is of particular interest in relation to induc-

[1]The figures appearing in parentheses pertain to the bibliography appended to this paper.

tion hardening because it shows very clearly the importance of heating to high temperatures to obtain rapid austenitization. For example, the complete transformation of fine pearlite to austenite occurs in 100 seconds at 730 degrees Cent. (1345 degrees Fahr.), but occurs in only 9 seconds at 765 degrees Cent. (1410 degrees Fahr.). The complete disappearance of residual carbides and the subsequent homogenization require greatly increased lengths of time—estimated by them to be more than 10,000 seconds at 765 degrees Cent. (1410 degrees Fahr.). Thus a steel may be effectively hardened in a short period, although marked inhomogeneities still exist within it. The results published by Lauderdale and Harder are also of interest in this connection: they show that at 870 degrees Cent. (1600 degrees Fahr.) less than 2 seconds is necessary to obtain full hardness in small quenched samples of eutectoid steel (5).

The rates of transformation given above are fast when considered from the point of view of conventional heat treating practice, but are slow when considered from the point of view of obtainable rates of heating by induction methods. For example, Sherman has surface-hardened steel samples in a total heating time of 0.6 second (6), and the authors have hardened small samples of 0.50 per cent carbon steel in less than 0.3 second total heating time (the heating rate in this instance probably exceeded 3000 degrees Cent. per second). These results signify that the formation of austenite is exceedingly rapid at temperatures near 1000 degrees Cent. (1830 degrees Fahr.). The work by Martin and Wiley on induction hardening of plain carbon steels illustrates this point (7). Their results indicate that for a heating rate of about 100 degrees Cent. (180 degrees Fahr.) per second a surface temperature of 900 degrees Cent. (1650 degrees Fahr.) must be reached in order to obtain complete transformation to austenite at the surface of a normalized 1050 sample—in contrast to a temperature of about 770 degrees Cent. (1420 degrees Fahr.) by conventional methods of heat treatment.

Effect of Initial Microstructure—The austenitizing reaction is influenced by the initial microstructure as well as by the reaction temperature. Roberts and Mehl have shown that both the rate of nucleation and rate of growth are structure-sensitive; specifically showing that as the pearlite spacing is decreased, the rate of nucleation and rate of growth increase (1).

In induction hardening the initial structure is of particular importance because of the rapid heating which may involve rates over

2000 degrees Cent. (3600 degrees Fahr.) per second. It has been observed that an initial sorbitic structure is ideal for induction hardening of medium carbon steels (7), (8), (9). The reason is clear—the cementite is more uniformly distributed in the sorbitic sample; therefore, the diffusion path of carbon is shorter than in a normalized structure with free ferrite and pearlite.

The initial structure is of less significance in the high carbon steels because of the absence of free ferrite; therefore, these steels may be preferred to medium carbon steels for some induction hardening applications (7).

Effect of Composition—The only known information on the effect of composition on the formation of austenite is that given for the effect of carbon (1), (5). As is to be expected the results indicate that either free ferrite (hypoeutectoid steel) or free cementite (hypereutectoid steel) prolongs the period for complete austenitization. In an induction-hardened sample the presence of undissolved ferrite is likely to have greater damaging effects than residual carbide; and, therefore, it is of greater necessity to obtain complete austenitization in hypoeutectoid steels.

It is obvious that additional quantitative information is needed on the effect of alloying elements on the formation of austenite since many alloy steels are available for induction hardening applications; and at present we have no rules to guide us in their selection.

Formation of Martensite

Mechanism—The formation of martensite from austenite differs from the austenite→bainite and austenite→pearlite transformations in that the reaction does not involve nucleation and growth, but is a shear-type of reaction which is independent of the diffusion of carbon. Austenite when cooled rapidly, to prevent the formation of pearlite and bainite, will start to transform to martensite at the Ar″ or M_s temperature and continue to form on cooling down to the M_f point, the temperature at which the transformation is essentially completed (10), (11), (12).

Effect of Alloy Additions—The role of alloys in the hardening of steels is primarily one of delaying the start of the austenite→pearlite transformation. As a result, a slower quench will still result in the formation of martensite, and thus permit hardening of larger masses of metal. That is, the alloying elements, in general,

lower the critical cooling velocity for the formation of martensite, and increase the hardenability of the steel.

The classical work of Grossmann in calculating hardenability from chemical composition was a great step forward (13) in this field. His work has been extended to a wide range of alloying elements (14), (15); thus enabling the calculation of the hardenability of a steel of almost any composition.

Effect of Initial Structure—The rapid heating by induction does not in general allow time for homogenization of austenite. Consequently, it is to be expected that the austenite formed during induction heating generally contains carbon concentration gradients. In addition undissolved carbide and ferrite are frequently present which will likewise have their effect on the hardenability of the steel. But the question is, how will these inhomogeneities in structure alter the transformation characteristics?

Undissolved carbides when present in considerable quantity materially reduce the hardenability of steels (16), (17). Carbon concentration gradients have little effect upon hardenability of high carbon steels (17). In contrast, the medium carbon steels appear to be influenced greatly by carbon concentration gradients and chemical segregation, as several investigators have shown that the hardenability is affected by holding time at the austenitizing temperature (18), (19). However, the hardenability, as measured by the depth of penetration on a Jominy bar, may not give the complete story as to the effects of incomplete austenitizing during induction hardening. For example, hardenability has little significance in the surface hardening of a half-inch diameter shaft; nevertheless, localized inhomogeneities in the sample prior to quenching may greatly alter the properties of the surface layer. In particular, free ferrite in the hardened case may be expected to lower the fatigue properties; undissolved carbide, on the other hand, probably would not have any great effect in a martensitic matrix.

It is difficult to present a coherent picture at the present time because of the lack of quantitative data. Roughly it appears that in steels above about 0.6 per cent carbon the effect of inhomogeneities in the austenite are of little moment in surface hardening, although they may be of significance in deep hardening. For lower carbon steels, however, the structural irregularities may have damaging effects even in shallow hardened parts.

Untransformed Austenite—The M_s and M_f points are influenced

by the composition and the initial nature of the austenite. These temperatures are lowered by increasing carbon content and by most alloy additions (20), (21). As a result, high carbon or highly alloyed steels will have a considerable quantity of untransformed austenite in the quenched condition (22), (23).

Homogenization also tends to stabilize the austenite; consequently, induction-hardened samples with undissolved carbides and carbon concentration gradients may contain more martensite than a similar steel quenched after longer austenitizing treatments (7). The presence of untransformed austenite in any hardened steel would result in a lower hardness. Actually, induction-hardened samples have been observed to be harder than furnace-hardened samples of the same steel and the presence of retained austenite may explain the difference (7), (8), (9), although internal stresses resulting from the increase in volume due to the formation of martensite may also account for part of the higher hardness of surface-hardened steels (7), (24).

Object of Investigations

Technical knowledge of the metallurgy of induction heat treatment of steels has lagged commercial developments in the applications of induction heating methods. The present paper is an account of a series of experiments designed to help establish a better understanding of the metallurgical factors which are of importance in induction hardening of steel. In particular, the object is to study the effect of alloying elements and microstructure on the austenitizing characteristics of medium carbon steel and to attempt to correlate the results with the induction hardening characteristics. In addition, attention has been given to the study of the effect of retained austenite and internal stresses on the hardness of hardened steel.

Experimental Procedure and Results

Materials—Since medium carbon steels are favored in many induction hardening applications it was decided to study a series of 0.50 per cent carbon steels. The analyses of the five steels used are given in Table I. The S.A.E. 1350 steel was prepared in a 50-pound laboratory induction furnace, cast into molds 2 inches in diameter, and hot-forged to 1-inch diameter bars. The S.A.E. 1050, 2350,

Table I
Chemical Analyses and Critical Temperatures

Type	C	Mn	Si	Ni	Cr	Mo	Ac_1	Ac_3
1050	0.47	0.73	0.11	....	....	....	725	765
1350	0.50	1.46	0.43	....	....	....	710	750
2350	0.52	0.32	tr.	3.44	0.28	<0.1	690	725
4160	0.60	0.90	0.03	tr.	1.15	0.22	750	775
9255	0.52	0.91	1.88	....	....	....	770	800

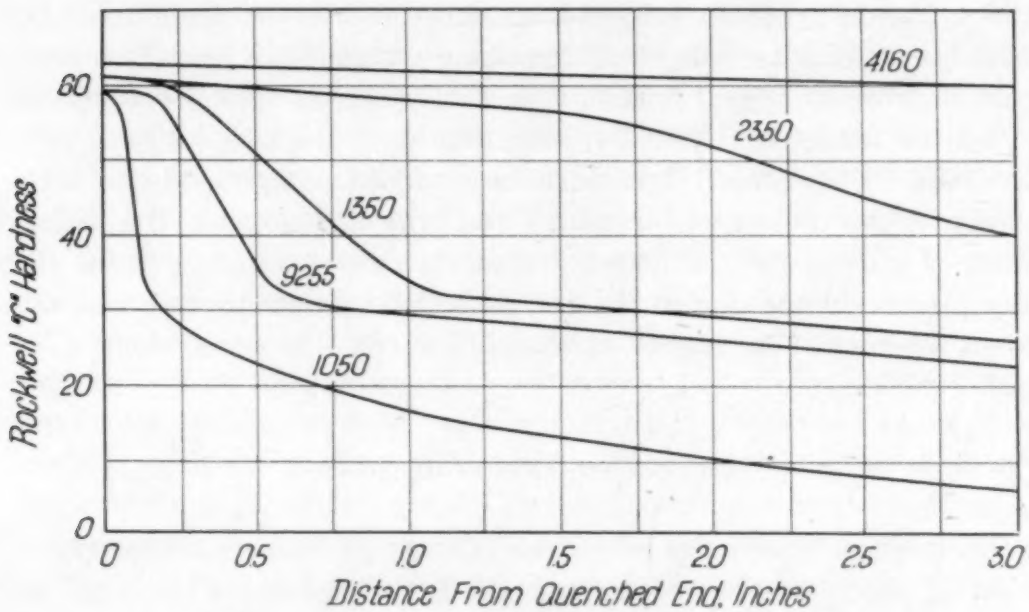

Fig. 1—Jominy Hardenability Results Obtained Upon Five Steels. The heat treating information is given in Table II.

4160,[2] and NE 9255 steels were purchased from commercial sources as 1-inch diameter bars.

Critical Temperatures—Critical Temperature information for the steels was obtained from dilatometric studies. Small dilatometric samples were machined from the five steels and curves obtained for a heating rate of 200 degrees Cent. (360 degrees Fahr.) per hour. The Ac_1 and Ac_3 temperatures given in Table I were obtained from these curves. It is noted that the S.A.E. 1350 and 2350 steels transform to austenite at temperatures below that of the plain carbon 1050 steel; while the S.A.E. 4160 and NE 9255 steels transforms to austenite at higher temperatures.

Hardenability—The Jominy end-quenching technique was followed for determining the hardenability of the five steels (25). The results given in Fig. 1 and Table II show a wide variation in their

[2]This steel was purchased as S.A.E. 4150, but analysis showed it to contain 0.60 per cent carbon; accordingly, the steel will be referred to herein as 4160.

Table II
Hardenability of the Steels (Jominy Test)

Steel	Quenching Temperature Degrees Cent.	Austenite Grain Size A.S.T.M.	Depth to 50 Rc 1/16 Inch	Hardness Quenched End Rc
1050	800	7	2	59
1350	800	6–7	8	61
2350	800	6–7	34	62
4160	850	6–7	>48	64
9255	870	6	5	59

hardenability characteristics. The S.A.E. 1050, 1350, and NE 9255 steels are shallow to medium hardening and the S.A.E. 2350 and 4160 steels are deep hardening.

The wide spread in the hardenability and critical temperatures of the steels should afford a broad comparison with the austenitizing and induction hardening results.

Austenitizing Characteristics

In contrast to the extensive information on the effect of alloying elements on hardenability of steel very little is known about the effect of alloying additions on the rate of formation of austenite from aggregates of cementite and ferrite.

The most effective method of studying the influence of alloying elements on austenitizing appears to be the isothermal reaction method which was used by Roberts and Mehl in their study of the mechanism and rate of formation of austenite in plain carbon steels (1). Briefly, the method consists of determining the isothermal reaction curves at several temperatures and plotting a time-temperature isotransformation diagram (TTT diagram) from the data. A comparison of the diagrams for the different steels should reveal the effect of the various elements on the austenitizing reaction.

Procedure—The general experimental technique consisted of heating small thin samples (¼-inch diameter by 0.035 inch) in a molten lead bath at a constant temperature for various intervals of time, and then estimating the extent of the reaction in the quenched samples by following the corresponding increase in hardness.

To obtain consistent results it was found necessary to hold the specimen thickness to close tolerances, and for this program the samples were cut to 0.035 ± 0.003 inch. The time to heat these specimens to the bath temperature was estimated to vary from 2 to 3

seconds depending upon the temperature. The heating time was subtracted from the total immersion time to obtain the isothermal reaction time.

The total immersion time was measured with a stop watch. While this method of measurement may introduce error, particularly in the shorter heating periods, close checks were obtained on repeated samples and it is felt that the values reported are accurate.

The furnace temperature was controlled with a photo-electric-

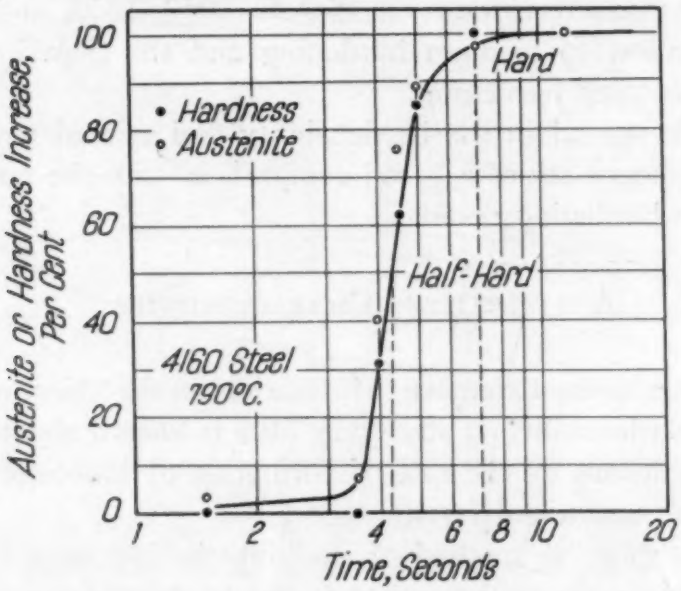

Fig. 2—Typical Isothermal Reaction Curve for the Formation of Austenite. Comparison of the microscopic and hardness methods of measuring the extent of the reaction. The half-hard and hard values are indicated.

type controller to better than ± 1 degree Cent., and the bath temperature measured with a platinum : platinum-rhodium thermocouple.

Error due to absorption of heat during transformation is considered to be negligible because of the small volume of the samples and the high heat capacity and conductivity of the molten lead bath.

After isothermal reaction the quenched samples were cleaned with emery paper and hardness readings (Rockwell "A" scale) were made. About seven readings were found to yield a satisfactory average value.

Roberts and Mehl measured the extent of the austenitizing reaction by microscopic methods; but also found that the hardness increase resulting from transformation indicated the approximate extent of the austenite reaction. The microscopic method is a time-con-

suming operation which was not considered justified in the present investigation; and, therefore, it was decided to measure the extent of the reaction by the hardness increase.

One series of samples was checked by both methods and the resulting curve is given in Fig. 2. The agreement between the two methods is good for this particular steel; however, the results of Roberts and Mehl indicate that the hardness curve generally lags behind the microscopically determined curve up to 10 or 15 per cent reaction, and the hardness reaches an apparent maximum when but 85 to 90 per cent of the pearlite (in a eutectoid steel) has transformed to austenite (1).

Another reason for adopting the hardness method is that from an induction hardening viewpoint the time to obtain maximum hardness is of greater practical use than the time to dissolve all carbide; on the other hand, the complete elimination of free ferrite may be desired in induction-hardened samples, but the information on this subject is as yet incomplete.

Initial Microstructure—The austenitizing characteristics were measured on furnace-cooled samples. The microstructures of the steels in the initial condition before heating in the lead bath are given in Figs. 3 and 4. Note that there is more free ferrite in the S.A.E. 1050 steel, and also that the 2350 and S.A.E. 4160 steels are close to the eutectoid composition. The S.A.E. 1350 and 9255 steels have intermediate positions between these two groups.

Alloy additions alter the structure by virtue of the shift they cause in the eutectoid composition, and also raise or lower the critical temperature range. Both of these factors will be shown to have an important bearing on the austenitizing characteristics of the steels.

TTT Diagrams—The basis of the TTT diagram is the isothermal reaction curve, an example of which is given in Fig. 2. The "half-hard" and "hard" values are taken from the isothermal reaction curve and plotted versus the reaction-bath temperature, as shown in Fig. 5 for a series of temperatures above the Ac_1 temperature. The pearlite to austenite reaction (the S.A.E. 4160 steel is eutectoid) proceeds very slowly at temperatures near the Ac_1, but the rate increases rapidly with increase in temperature; near 825 degrees Cent. (1520 degrees Fahr.) the reaction is completed in less than 1 second.

The TTT diagrams for the other four steels are given in Figs. 6, 7, 8, and 12. For comparison the results have been replotted in Figs. 9 and 10. The half-hard curves in Fig. 9 are seen to be ar-

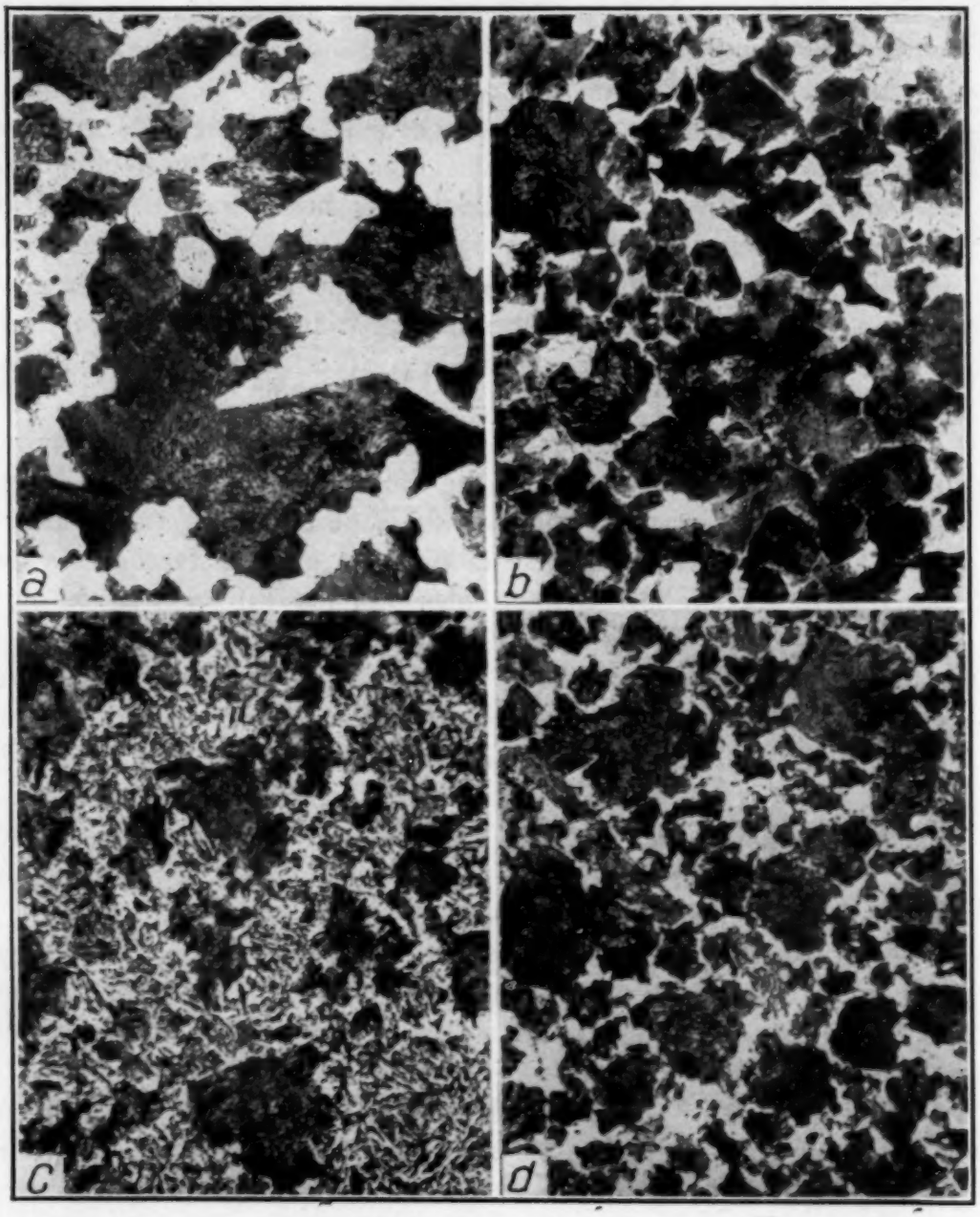

Fig. 3—Initial Microstructure of the Furnace-Cooled Steels. × 250. (a) S.A.E. 1050; (b) S.A.E. 1350; (c) S.A.E. 2350; (d) S.A.E. 9255.

ranged in the order of increasing critical temperatures (the Ac_1 and Ac_3 temperatures are listed in Table I) with 2350 exhibiting the fastest austenitizing rate and S.A.E. 9255 the slowest. Two alloy steels, S.A.E. 2350 and 1350, are half-hardened in shorter times than the plain carbon S.A.E. 1050 steel, while the S.A.E. 4160 and 9255 steels are sluggish in comparison to the S.A.E. 1050 steel.

Surprising results are to be found in Fig. 10 where the full-hard curves are compared. The four alloy steels are hardened in a shorter time at temperatures above 800 degrees Cent. (1470 degrees Fahr.) than the furnace-cooled S.A.E. 1050 steel. Except for the 1050 steel the order of the curves is consistent with the order of increasing critical temperatures.

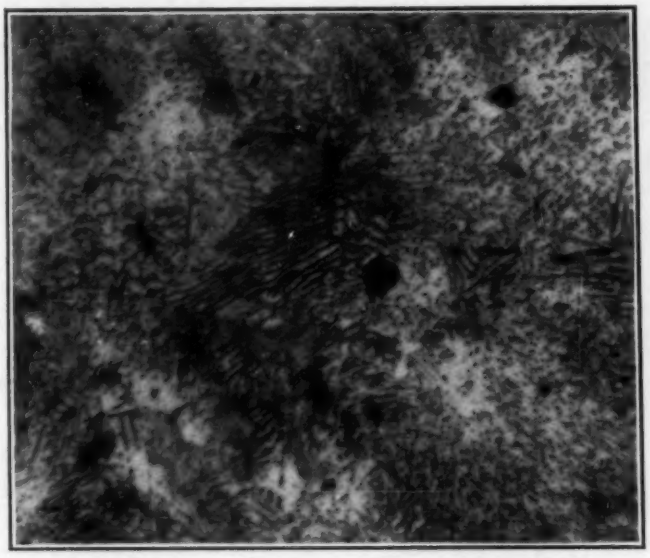

Fig. 4—Initial Microstructure of the Furnace-Cooled S.A.E. 4160 Steel. × 1000.

The sluggishness of the S.A.E. 1050 steel in hardening is likely due to the large amounts of free ferrite present in the S.A.E. 1050 steel in contrast to the alloy steels (Figs. 3 and 4).

Effect of Initial Microstructure—The initial microstructure of the 1050 steel was altered by quenching from 800 degrees Cent. (1470 degrees Fahr.) and tempering for 1 hour at 500 degrees Cent. (930 degrees Fahr.). The resulting sorbitic structure is compared with the furnace-cooled microstructure in Fig. 11. It is obvious that because of uniform carbide distribution the time to harden will be much less for the sorbitic samples. The austenitizing results in Fig. 12 show that the sorbitic samples do harden much more rapidly than the same steel with a furnace-cooled initial microstructure. In fact the sorbitic S.A.E. 1050 steel has hardening characteristics very similar to the S.A.E. 1350 steel in Fig. 10, as might be expected from the results in Fig. 9.

The sorbitic samples of S.A.E. 2350 steel have TTT curves very close to those of the furnace-cooled material shown in Fig. 7. Examination of Fig. 3c shows that the cementite is evenly distributed

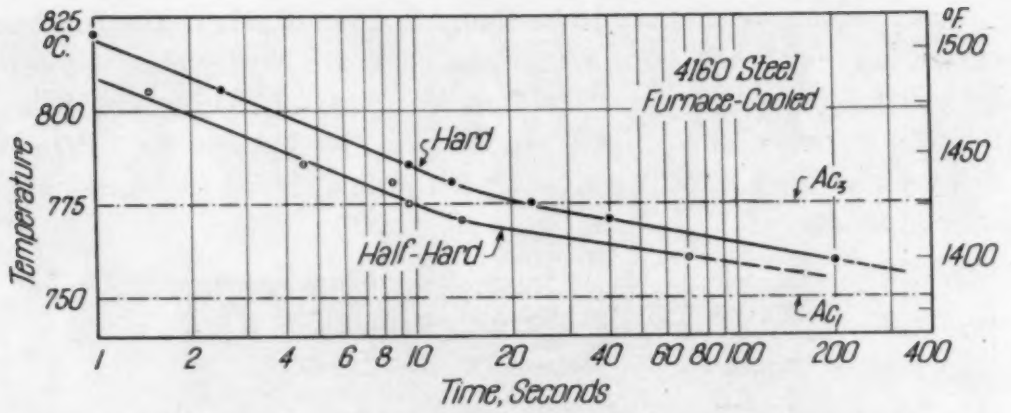

Fig. 5—Austenitizing TTT Diagram for S.A.E. 4160 Steel. The initial structure is given in Fig. 4.

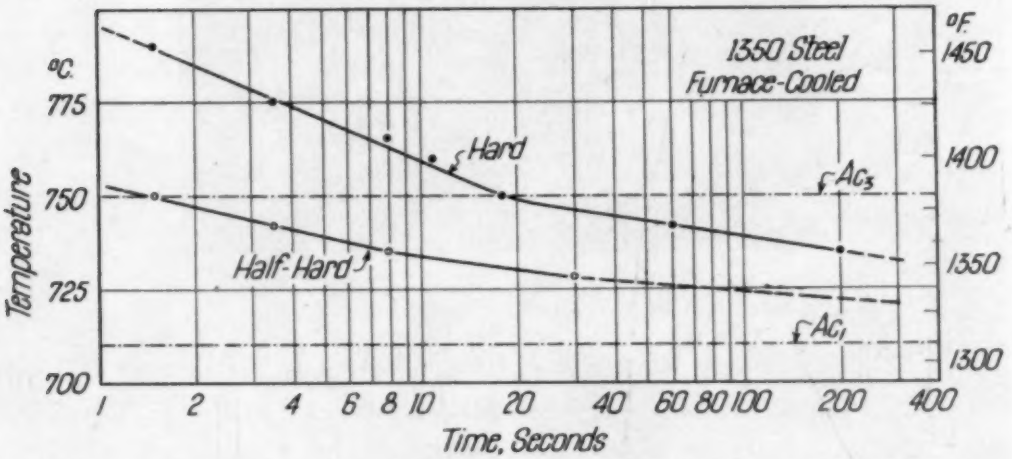

Fig. 6—Austenitizing TTT Diagram for S.A.E. 1350 Steel. The initial structure is given in Fig. 3b.

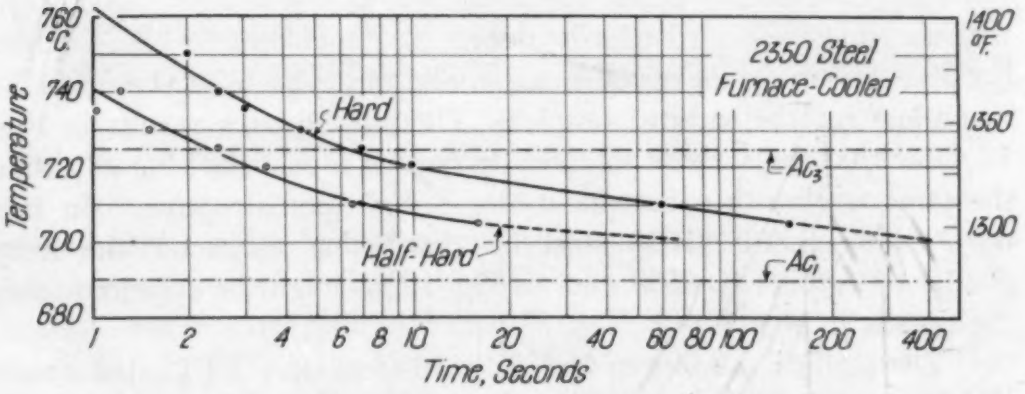

Fig. 7—Austenitizing TTT Diagram for S.A.E. 2350 Steel. The initial structure is given in Fig. 3c.

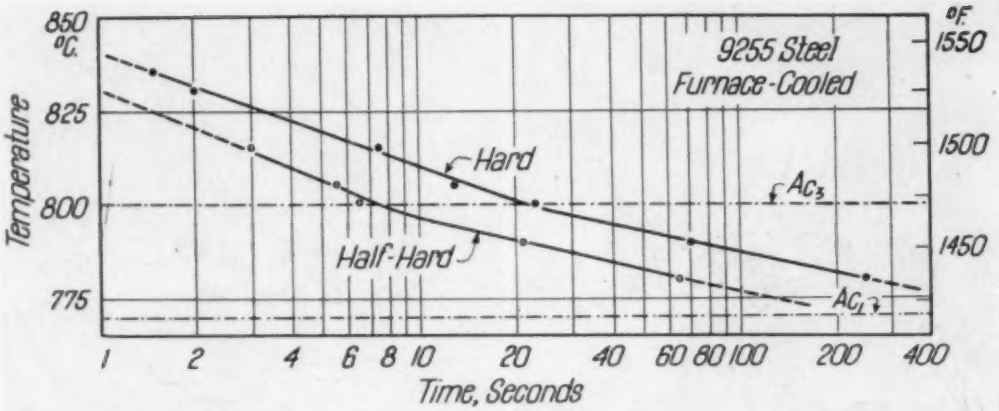

Fig. 8—Austenitizing TTT Diagram for S.A.E. 9255 Steel. The initial structure is given in Fig. 3d.

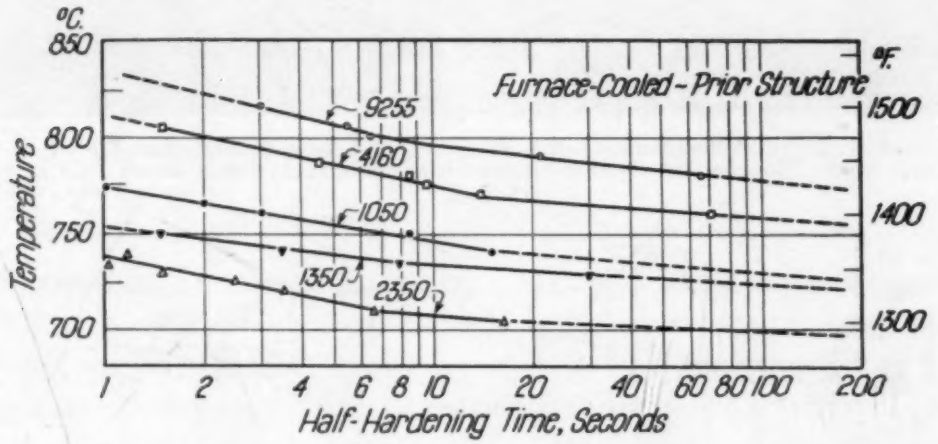

Fig. 9—Comparison of the Half-Hard TTT Curves for Five Steels.

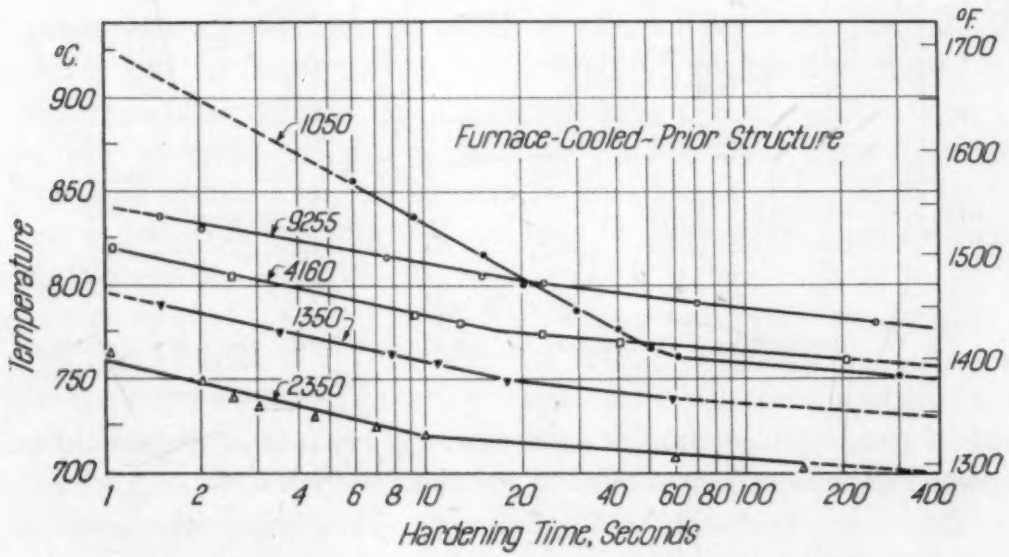

Fig. 10—Comparison of the Hard TTT Curves for Five Steels.

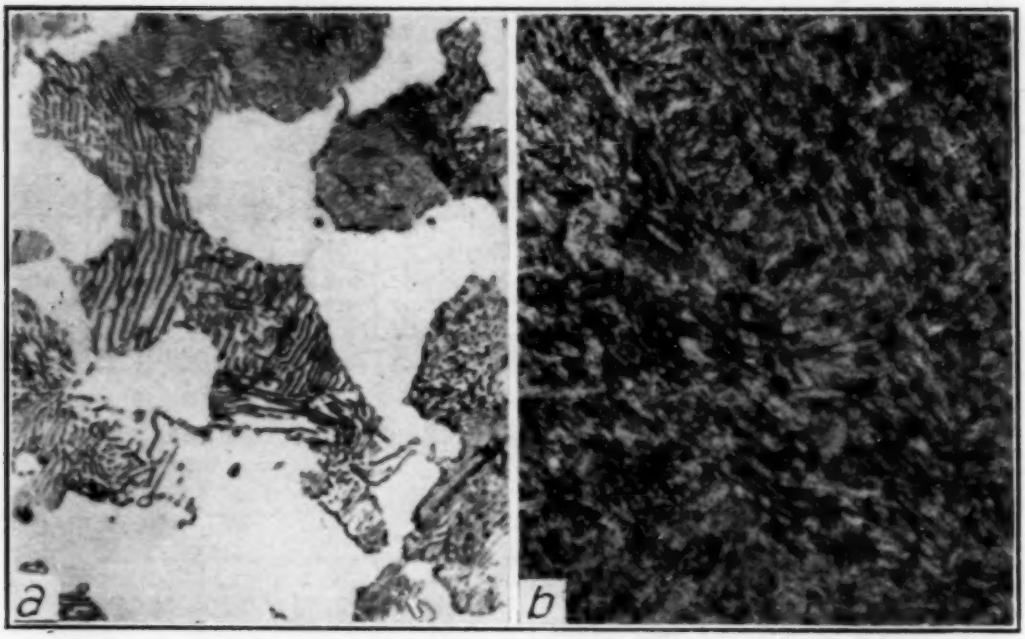

Fig. 11—Microstructure of the Furnace-Cooled and Sorbitic Samples of S.A.E. 1050 Steel. The effect of these structures on austenitizing rates is shown in Fig. 12.

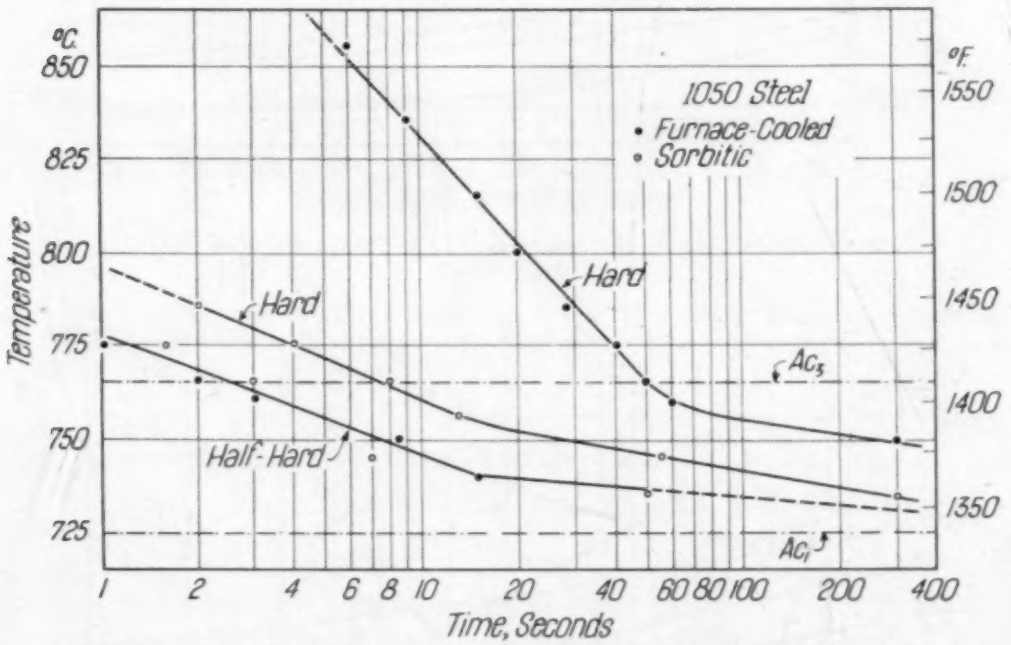

Fig. 12—Austenitizing TTT Diagram for S.A.E. 1050 Steel. The initial structures are compared in Fig. 11.

in the furnace-cooled condition; accordingly, no advantage should be expected to result from producing a sorbitic structure in such a steel.

Effect of Austenite Homogenization—Inhomogeneities exist in austenite when it is first formed from ferrite-cementite aggregates.

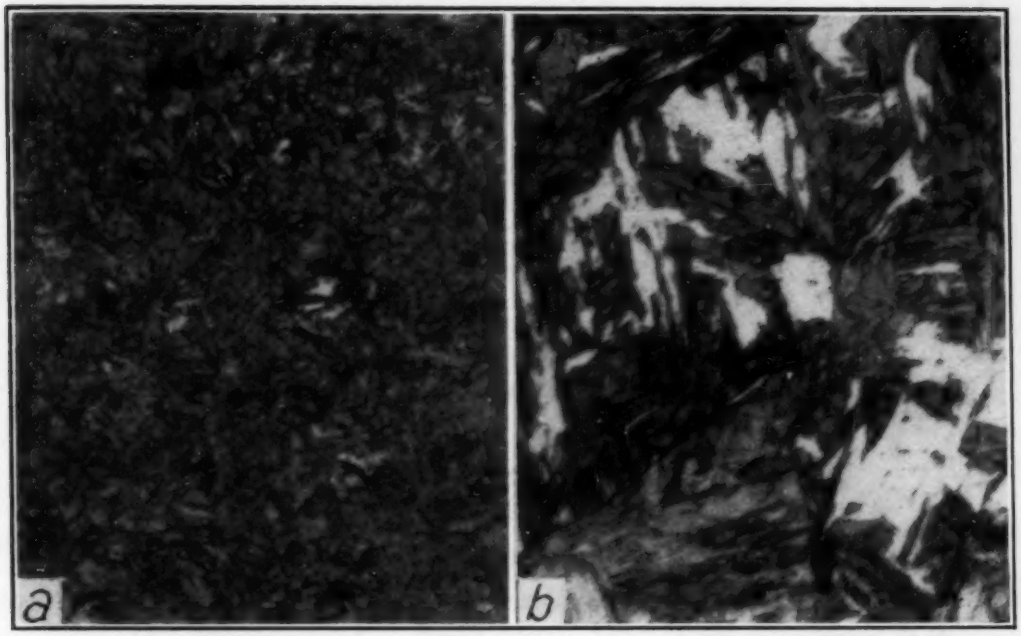

Fig. 13—Comparison of Microstructure of Quenched Samples of S.A.E. 2350 Steel; × 1000. Fig. 13a—Held 3 seconds at 850 degrees Cent.; 61 Rc hardness. Fig. 13b—Held 3 hours at 850 degrees Cent.; 59 Rc hardness.

This condition may be removed by prolonged heating which eliminates the structural irregularities such as carbon concentration gradients. However, the homogeneous austenite when quenched does not always transform completely to martensite unless the cooling is continued to very low temperatures. For this reason, the steels with undissolved carbides, in some instances, may contain more martensite (less untransformed austenite) than similar steels quenched after prolonged austenitizing treatments. The results for the S.A.E. 2350 steel in Fig. 13 illustrate such an example. The specimen heated for 3 hours at 850 degrees Cent. (1560 degrees Fahr.) has more untransformed austenite and was softer than the specimen austenitized for 3 seconds at the same temperature. This was also found to be the case for the S.A.E. 4160 steel and may be expected to hold for practically all steels.

In view of the slight drop in hardness with prolonged heating it may be desirable not to homogenize during austenitizing (unless a low temperature transformation treatment is also given) when maximum hardness is sought through heat treatment. The rule to heat "1 hour per inch" may be found to be inadequate for many heat treating problems. The safest procedure might be to determine the time-temperature conditions for optimum properties.

Table III
Heating Time and Hardness Penetration Data for the Five Steels

Temp., °C.	720		730		750		790		825		900		975	
	Heating Time and Penetration†													
Steel*	Sec.	Mils	Sec.	Mils	Sec.	Mils	Sec.	Mils	Sec.	Mils	Sec.	Mils	Sec.	Mils
1050	...	.....	...	.....	6.4	0.025	8.4	0.060	9.6	0.075	11.7	0.095	15.4	0.135
1350	...	.....	...	.....	6.8	0.030	8.5	0.090	10.8	0.145	12.4	0.310	....	.....
2350	3.6	0.015	6.4	0.040	6.9	0.060	8.2	0.130	9.6	0.190	14.6	0.375	16.0	0.375‡
4160	...	.....	...	.....	3.3	0.000	8.7	0.030	10.2	0.060	13.2	0.280	16.0	0.375‡
9255	...	.....	...	.....	5.4	0.000	7.3	0.000	8.8	0.025	10.9	0.080	15.6	0.220

*All specimens furnace-cooled prior to induction heating.
†Depth to 550 Knoop.
‡Hardened to center.

Induction Hardening Characteristics

The austenitizing results in the preceding section indicate that some alloy steels are transformed more readily to austenite on heating than the plain carbon S.A.E. 1050 steel. Accordingly, these same steels should have superior induction hardening characteristics, that is, they should be capable of being hardened with greater ease than the S.A.E. 1050 steel. Specimens of the same five steels previously studied (in the furnace-cooled condition) were induction-hardened to verify this assumption by checking the hardness penetration and the microstructure.

Procedure—The induction heating equipment and procedures used have been described in detail in a previous paper (*7*) and are reviewed here only briefly. A General Electric 15KW, 530,000-cycle electronic heater was used to heat the specimens which were 1 inch high and 7/8 inch in diameter. The samples were heated in a special fixture to the desired surface temperature and then quenched with a strong water spray.

A chromel-alumel thermocouple was spot welded to the surface of the sample and connected to a photoelectric temperature indicating instrument whose rate of response greatly exceeded the heating rates encountered in the specimens in all useful temperature ranges. In the induction heating operation the indicator was set for the desired temperature and when this value was reached, the primary heating circuit was opened and the quenching spray turned on simultaneously by the pressing of one push button. This was operated manually and the lag in human response is estimated to be less than 1/10 of a second or equivalent to approximately a 10 degree Cent. (20 degree Fahr.) rise in temperature. The heating time was recorded on an

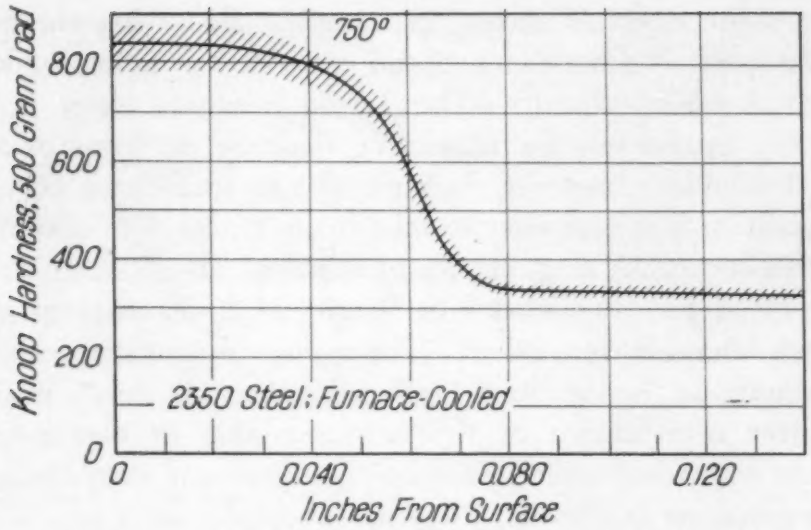

Fig. 14—Hardness Penetration Curve for Induction-Hardened S.A.E. 2350 Steel. The spread in Knoop hardness values is indicated.

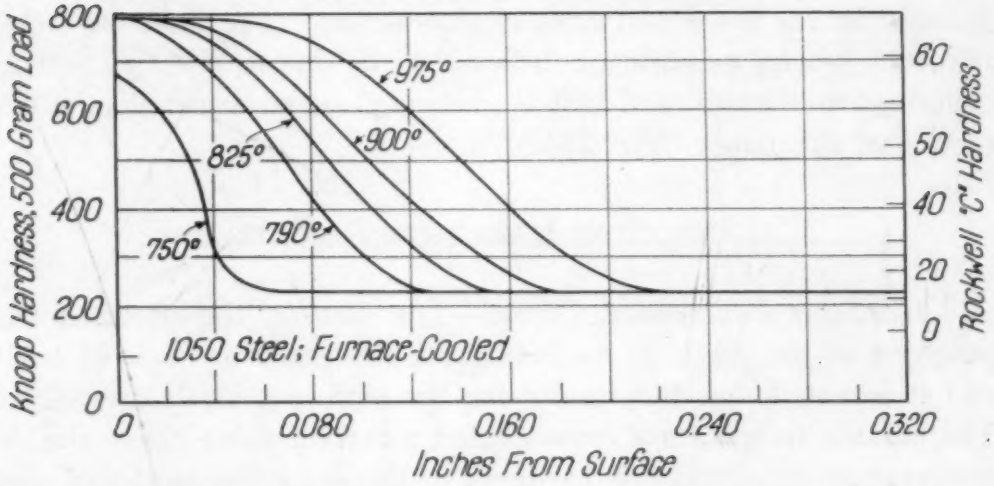

Fig. 15—Hardness Penetration Curves for S.A.E. 1050 Steel. The specimens were induction-heated to the designated temperatures (degrees Cent.) and quenched.

electronic timer directly connected to the on-off push buttons of the induction heater.

The heating times for the steels are listed in Table III. Note that the rate of heating is rapid up to the Curie temperature, but slows down considerably beyond this value.

The induction-hardened specimens were sectioned at mid-height, using a high speed cutting wheel which was well flooded with water during the cutting operation. One-half of each specimen was chrome plated (to prevent rounding during metallographic preparation), mounted in bakelite and suitably polished. Hardness measurements, begun at the thermocouple weld location, were taken with a Tukon

tester at short intervals across the sample. Following these operations, the specimens were etched and examined metallographically.

Microhardness Results—The Tukon hardness tester is a very satisfactory instrument for measuring the case hardness of shallow hardened samples; however, because of the small area covered by the indentation it is necessary to take many readings to obtain a representative curve. A single series of readings may lead to erroneous results (7), and for this work 4 or 5 series of 20-50 readings at about 0.010-inch intervals were taken. The spread in microhardness which is encountered in the induction-hardened samples is shown in Fig. 14. For clearer presentation of the hardness data in the penetration curves the individual points have been omitted and only the average curve is given, as in Fig. 15.

The induction-hardened samples were slightly influenced by the bakelite mounting operation; consequently, the readings are a trifle lower than for the initial hardened condition. The lowering of the hardness during mounting is believed to be due to relief of internal compressive stresses, and will be discussed more in detail in a later section of this paper (Fig. 23).

Induction Hardening Results

Hardness Penetration Curves—The data on the effect of temperature on the depth of hardening are presented in Figs. 15 to 19, and it is noted that the results for the different steels are similar. The surface temperature must exceed a certain value above the Ac_3 temperature to obtain full hardness in the induction-hardened zone. A temperature of 740 degrees Cent. (1365 degrees Fahr.) is low enough to accomplish this in the 2350 steel (Fig. 17) in contrast to 900 degrees Cent. (1650 degrees Fahr.) for the 4160 steel (Fig 18). The S.A.E. 1350 steel (Fig. 16) must be heated to above 770 degrees Cent. (1420 degrees Fahr.), the 1050 steel (Fig. 15) to 790 degrees Cent. (1455 degrees Fahr.), and the 9255 steel (Fig. 19) to 850 degrees Cent. (1560 degrees Fahr.) in order to obtain full surface hardness.

The S.A.E. 2350 and 1350 steels with lower critical temperatures (Table I) and less free ferrite (Fig. 3) than the 1050 steel are capable of being induction-hardened at lower temperatures (Figs. 15, 16, and 17). This is in agreement with the austenitizing results on these same steels (Figs. 9 and 10).

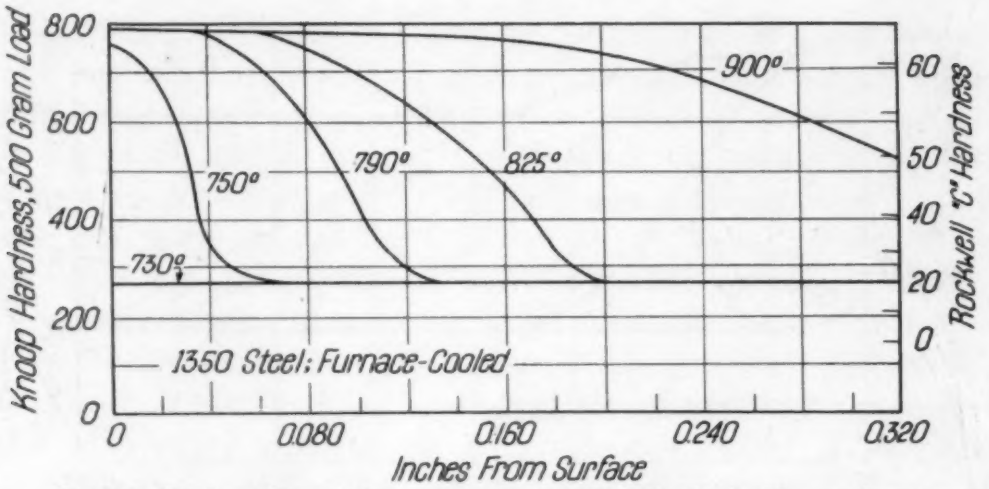

Fig. 16—Hardness Penetration Curves for S.A.E. 1350 Steel. The specimens were induction-heated to the designated temperatures (degrees Cent.) and quenched.

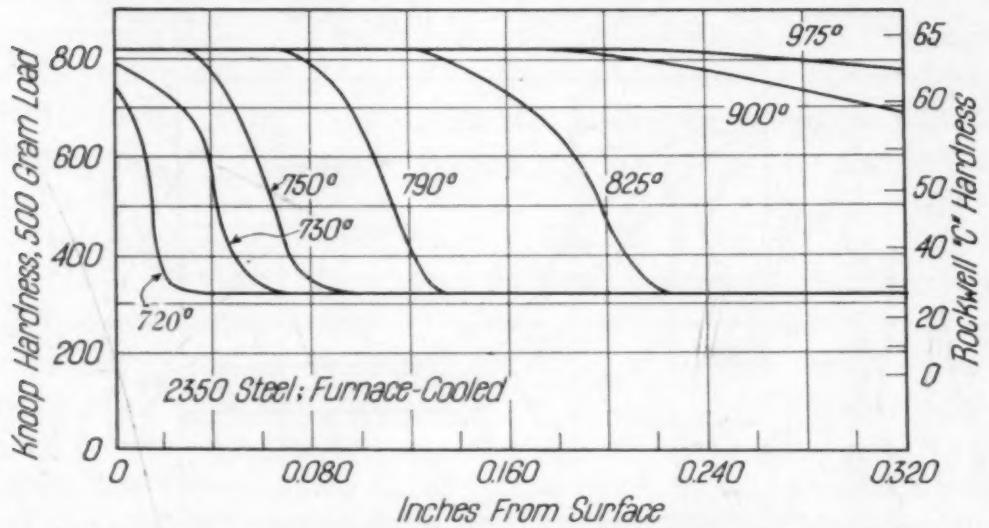

Fig. 17—Hardness Penetration Curves for S.A.E. 2350 Steel. The specimens were induction-heated to the designated temperatures (degrees Cent.) and quenched.

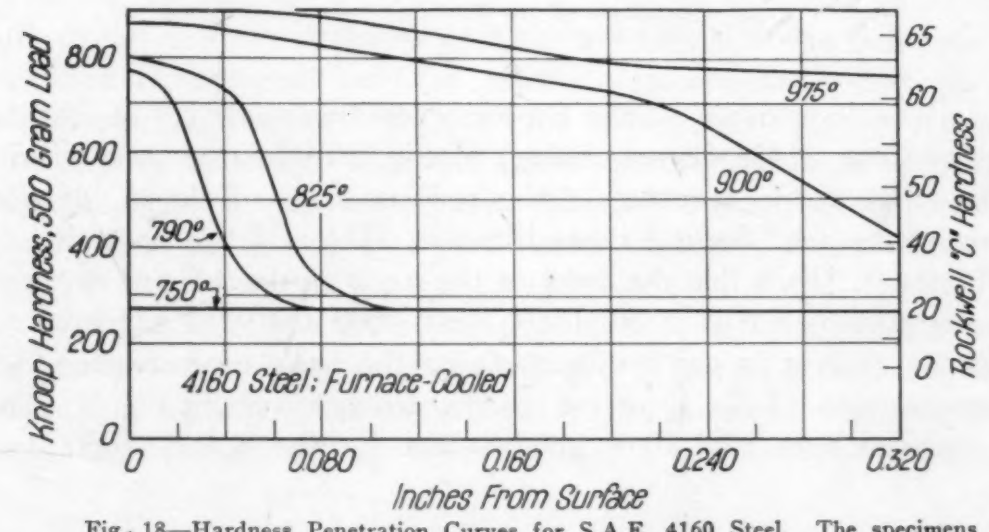

Fig. 18—Hardness Penetration Curves for S.A.E. 4160 Steel. The specimens were induction-heated to the designated temperatures (degrees Cent.) and quenched.

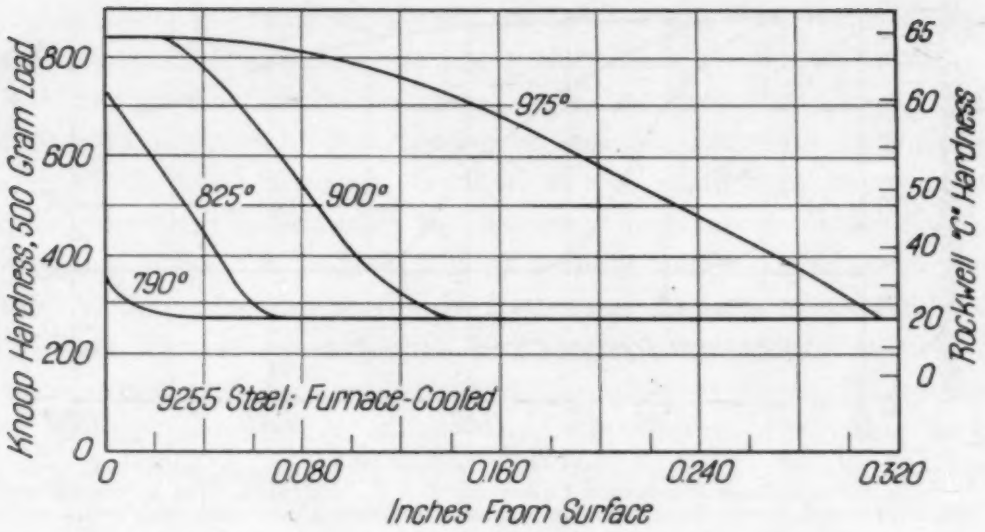

Fig. 19—Hardness Penetration Curves for S.A.E. 9255 Steel. The specimens were induction-heated to the designated temperatures (degrees Cent.) and quenched.

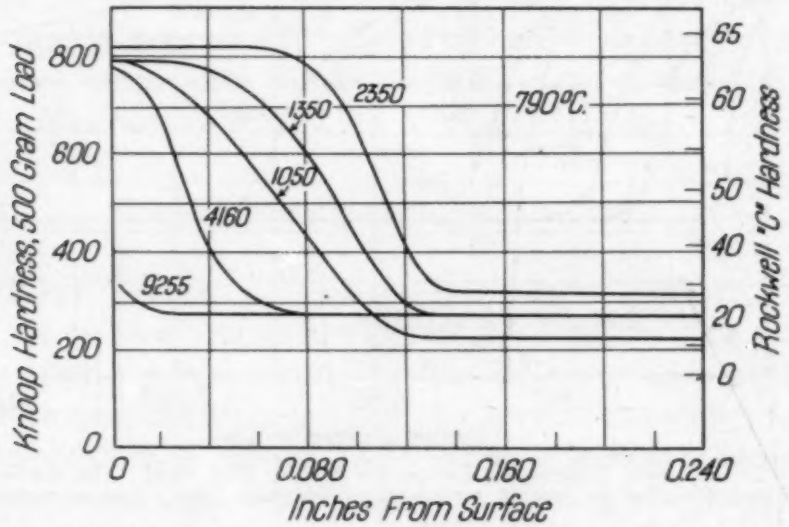

Fig. 20—Comparison of the Hardness Penetration Curves for Five Steels Induction-Heated to 790 Degrees Cent.

The comparison of the hardness penetration curves at 790 degrees Cent. (1455 degrees Fahr.) in Fig. 20 illustrates very clearly the difference in response of the various steels to almost identical treatments (the heating times listed in Table III are only slightly different). Note that the order of the steels on the basis of depth of penetration is S.A.E. 2350, 1350, 1050, 4160 and 9255 which corresponds exactly to the arrangement on the basis of increasing Ac_3 temperatures (Table I) or the half-hardening curves in Fig. 9. The agreement with Fig. 10 is good except for the S.A.E. 1050 steel

which appears to be out of order. However, the slight discrepancy is perhaps due to errors in the method of measuring hardness. The austenitizing curves are based on Rockwell readings and the penetration curves on Knoop values. The Rockwell "C" value on the 790 degree Cent. (1455 degree Fahr.) sample was found to be slightly lower than for the sample heated to 825 degrees Cent. (60 to 63 RC) in contrast to no difference on the Knoop hardness curve. Examination of the structure revealed some free ferrite in the hardened zone of the 790 degree Cent. (1455 degree Fahr.) sample. Apparently, the Rockwell measurement gives a more representative hardness value of a mixed structure of martensite and ferrite; although in Fig. 2 it was shown that even the Rockwell test does not indicate the presence of small amounts of martensite or ferrite.

As the surface temperature is increased the penetration of heat into the sample is deeper until eventually the sample is heated through to the center (Table III). For these conditions the hardenability characteristics (Fig. 1) determine the depth of penetration. Thus, when heated to 975 degrees Cent. (1785 degrees Fahr.) the S.A.E. 2350 and 4160 steels are hardened to the center while the shallow hardening S.A.E. 1050 steel hardens only to 0.135 inch.

Microstructural Changes—The transformations in structure which occur during induction heating of the steel samples prior to hardening will follow, in general, the series pictured in Fig. 21. These photomicrographs show the structural changes in a laboratory-prepared 0.47 per cent carbon steel (designated as L 1050) which exhibited larger areas of free ferrite than the S.A.E. 1050 steel shown in Fig. 11. The transformation to austenite is initiated in the pearlite areas (actually at the ferrite-cementite interface) at a temperature above the Ac_1, and these areas continue to grow as the temperature is increased until the sample is transformed completely to austenite.

In a hypoeutectoid steel the pearlite areas are readily converted to austenite; however, diffusion of the carbon into the free ferrite areas is time-consuming at temperatures immediately above the Ac_1. Accordingly, high temperatures are required to austenitize such steels in the short heating period generally obtained during induction heating.

The S.A.E. 1050 steel is seen in Fig. 15 to be fully hardened at the surface when induction heated to 790 degrees Cent. (1455 degress Fahr.); although examination of the microstructure reveals slightly less free ferrite than for the sample in Fig. 21d. The presence

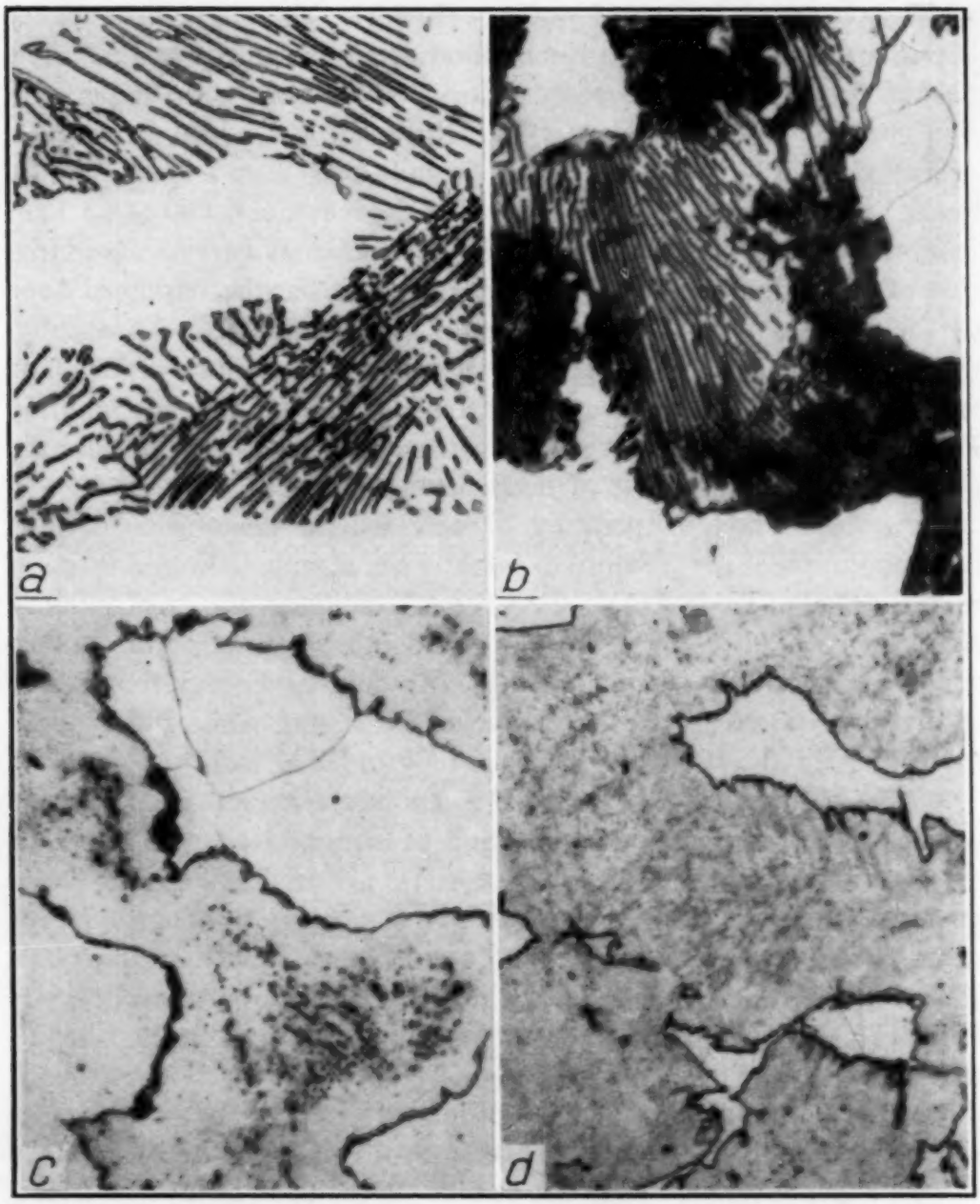

Fig. 21—Effect of Increasing Induction Heating Temperature on the Surface Microstructure of Steel L 1050; × 1000. Fig. 21a—Initial structure; Fig. 21b—Start of pearlite:austenite reaction, 735 degrees Cent. Fig. 21c—Transformation has covered pearlite areas, 750 degrees Cent. Fig. 21d—Austenite has grown into the ferrite areas, 800 degrees Cent.

of a small amount of ferrite may not be indicated by the hardness test, as previously discussed; however, it may impair the fatigue properties and should be avoided.[3] To dissolve the ferrite in the

[3]Examination of induction-hardened samples with a mixed structure of ferrite and martensite has shown that cracks originate in the soft ferrite in preference to hard martensite.

hardened zone of this 1050 steel it was necessary to heat to 825 degrees Cent. (1515 degrees Fahr.).

In eutectoid or hypereutectoid steels, elimination of free ferrite is not a problem, but solution of residual carbide may be. For example, the eutectoid S.A.E. 4160 steel when heated to 900 degrees Cent. (1650 degrees Fahr.) is seen in Fig. 22 to consist of residual carbides in a martensitic matrix. This structural constituent has been referred to as pearlite ghosts, or pseudomorphs of pearlite (1), (7), (26). It resembles pearlite in appearance but has the hardness

Fig. 22—Microstructure of S.A.E. 4160 Steel Induction-Heated to 900 Degrees Cent. The hardness of this structure is over 60 Rc.

of martensite, and is frequently observed in induction-hardened samples of high carbon steels. Little is known regarding the physical properties of such a structure in contrast to a homogeneous martensitic structure.

Effect of Internal Stresses—The surface induction hardening of a steel specimen results in a martensitic zone of high specific volume in direct union with a ferrite : cementite core of low specific volume. The martensitic zone is prevented from expanding to its natural volume because it is bonded to the soft core; as a result, high compressive stresses exist in the hardened layer. These internal compressive stresses may be of great usefulness in increasing the fatigue strength by partially neutralizing the external tensile stresses,

which unopposed might produce fatigue failures[4] (27), (28), (29). In addition, the internal stresses also appear to increase the hardness of the martensitic layer over that generally obtained by furnace hardening methods (7), (8), (24); however, because of the elusive nature of internal stresses it is difficult to prove this experimentally.

Crampton assumed that internal compressive stresses resist and tensile stresses assist the penetration of the hardness indenter (30). It follows that internal compressive stresses should raise and internal tensile stresses should lower the hardness. Likewise, relief of the compressive stresses should effect a decrease in hardness compared to an increase when residual tensile stresses are eliminated. Furthermore, he demonstrated this to be the case for a drawn brass tube with internal tensile stresses at the outside surface and internal compressive stresses at the inside surface. Similarly, it is to be expected that if shallow induction-hardened samples are slightly harder because of internal compressive stresses then stress-relief annealing should produce a drop in hardness. The deeply hardened samples are likely to have the surface in tension while the interior is in compression (31); therefore, stress-relief annealing should result in a slight increase in surface hardness.

To check this, 7/8-inch diameter samples of 1.1 per cent carbon drill rod were induction-hardened to various depths. The samples were sectioned and mounted in a low-temperature mounting material (60 to 80 degrees Cent.) to minimize tempering effects. The microhardness results in Fig. 23 show the shallow hardened sample to have slightly higher hardness than the deeply hardened sample before tempering, but of equal hardness after annealing at 100 degrees Cent. (210 degrees Fahr.) for 3 hours.[5] In view of the slight drop in hardness of the shallow induction-hardened sample when tempered it appears that the compressive stresses are responsible for at least part of the increase in hardness.

Effect of Internal Stresses on Cracking—The tensile stresses which exist at the surface in deeply hardened samples may exceed the tensile strength, in which event cracking occurs. The steels of high hardenability will harden to a greater depth more readily; and thus it is expected that they will be more susceptible to cracking than shallow hardening steels. However, other factors, such as tensile

[4]Shot peening, case carburizing, nitriding, flame hardening are a few of the other methods of introducing compressive stresses in the surface layer.

[5]The results in Figs. 15 to 19 do not show any variation with depth of hardening—probably due to tempering which occurred during mounting in bakelite.

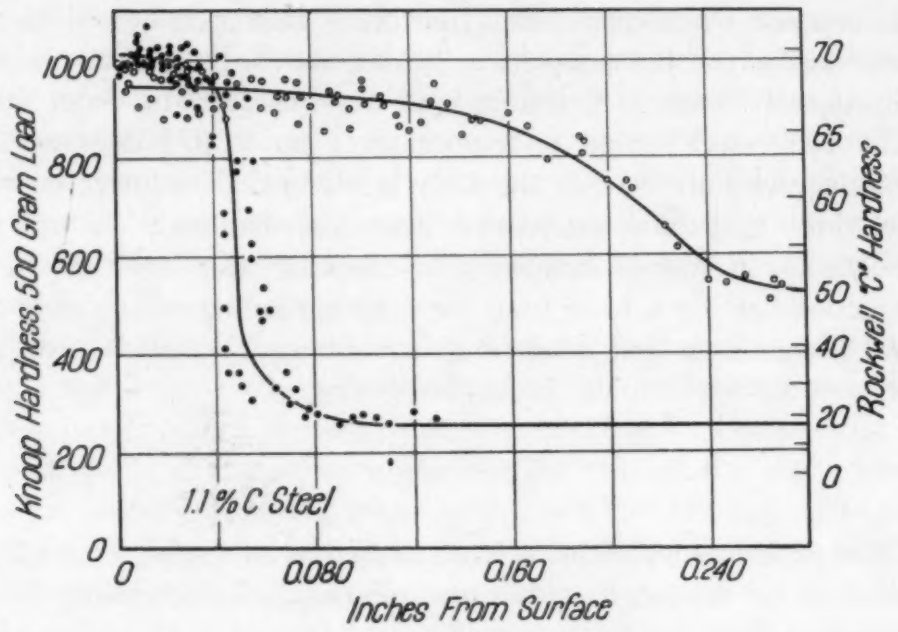

Fig. 23a

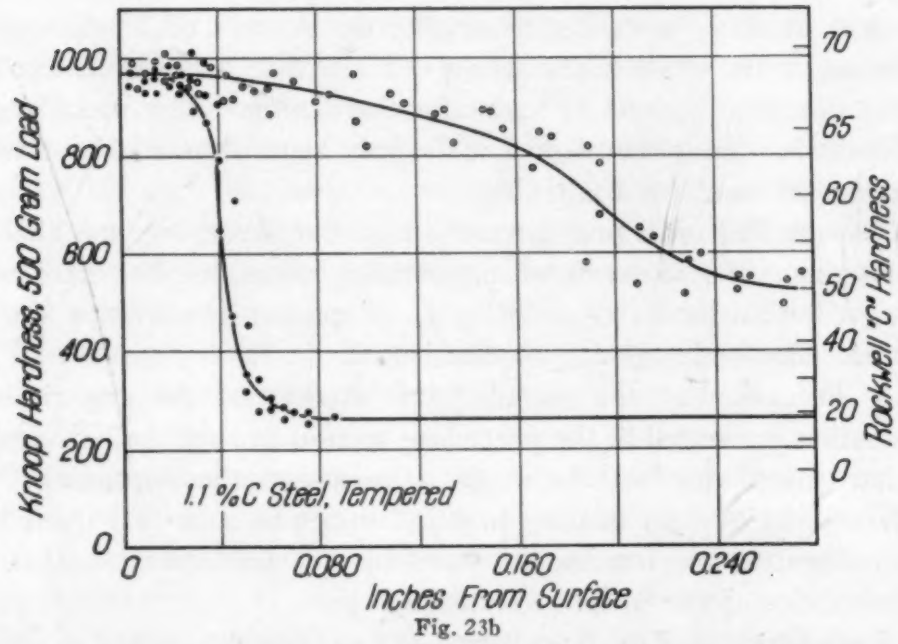

Fig. 23b

Fig. 23—Variation of Microhardness in Induction-Hardened Specimens of 1.1 Per Cent Carbon Steel with Different Depths of Penetration. Fig. 23a—As-hardened. Fig. 23b—After annealing at 100 degrees Cent. for 3 hours.

strength, cooling rate, etc., will influence the cracking characteristics; actually, the relative effects of the various factors are not known and additional information is needed to obtain a clear picture.

In the present experiments it was observed that the four alloy

steels cracked when quenched after being induction-heated to temperatures at which the sample was heated above the critical completely through to center. Cracking was not observed to occur in the S.A.E. 1050 steel heated to temperatures up to 975 degrees Cent. (1785 degrees Fahr.) or in the alloy steels heated to lower temperatures where a shallow martensitic zone was obtained. It was only in the deeply hardened samples that cracking was involved. Thus, if induction heating is to be used for deep hardening of any steel part, special precautions (oil quenching, for example) may be necessary to prevent cracks forming during hardening.

Discussion

The problems which arise from induction hardening of steel will, in general, be related to either the electrical characteristics of the induction oscillator or to the metallurgical factors. While the proper selection of the power and frequency ratings of the induction oscillator will in many instances determine the success of hardening applications, in the present discussion our attention will be centered on the metallurgical aspects of induction hardening. The metallurgical considerations are general and will apply regardless of the type of induction set used for hardening.

The metallurgical problems of induction hardening are likely to result from either incomplete austenitizing or incomplete hardening, although the austenitizing reaction is of greater importance since it must precede the hardening mechanism.

A discussion of the metallurgical aspects of the experimental information presented in the preceding section as they apply to induction hardening may best be made in terms of the hardening cycle which consists of (A) heating to form austenite, and (B) quenching to transform to martensite, followed by a brief description of the characteristics of the induction-hardened part.

Austenitizing—The transformation of ferrite: cementite aggregates to austenite will start above the Ac_1 temperature and continue on heating until the reaction is completed at some temperature above the Ac_3 temperature. The temperatures at which the reaction starts and ends will be influenced by the heating rate—accordingly an increase in the heating rate will shift the temperatures to higher values.

The general microstructural changes which occur on heating are illustrated in Fig. 21. The pearlite transforms first, followed by so-

lution of the ferrite, and lastly by homogenization of the austenite. Complete homogenization is seldom attained during induction heating because of the short heating period. Undissolved carbides and carbon concentration gradients are observed frequently in the induction-hardened specimens (as in Fig. 22), particularly in high carbon or alloy steels. These inhomogeneities in the austenite are likely to reduce the hardenability of the steel (1), and as will be shown later, may result in inferior properties.

The criterion for a satisfactory steel for surface induction hardening would be one which austenitizes readily on heating and which exhibits medium to deep hardening characteristics. Fortunately, this combination of characteristics was met by two of the steels studied in the present investigation; namely, the S.A.E. 1350 and 2350 steels. These steels were found to transform to austenite at lower temperatures than did the furnace-cooled samples of S.A.E. 1050 steel (Figs. 9, 10, and 20), and also to harden to a greater depth (Fig. 1). Two factors are responsible for the excellent austenitizing characteristics of the nickel and manganese steels: (a) The low critical temperatures (Table I); and (b) the uniform distribution of cementite (Fig. 3). The initial structures of the S.A.E. 4160 and 9255 steels are equally as favorable as the S.A.E. 1350 and the 2350 steels (Figs. 3 and 4), but their Ac_1 and Ac_3, as noted in Table I, are much higher. On the other hand, the critical temperatures of the 1050 steel are favorable, but the microstructure in the furnace-cooled condition with its large masses of free ferrite (Fig. 3) is far from ideal for induction hardening. Altering the carbide distribution by heat treatment greatly improves the austenitizing characteristics of this steel (Figs. 11 and 12).

Nickel and manganese lower the Ac_1-Ac_3 temperatures; whereas most of the other common alloying elements, such as molybdenum, vanadium, silicon, etc., raise the critical values. Accordingly, nickel and manganese may in the future be favored in steels to be used for induction hardening applications in preference to straight carbon or other alloy steels; however, the final selection will depend upon many factors: physical properties, machinability, availability, cost, etc.

The distribution of the cementite in the steel will vary with the composition and the heat treatment. All the common alloying elements shift the eutectoid composition to lower carbon content, and as a result, there is less free ferrite in alloy steels than exists in a similar plain carbon steel. This is illustrated in Figs. 3 and 4 where the

S.A.E. 1050 steel is seen to have larger amounts of free ferrite than the 0.50 per cent carbon alloy steels. The structure is altered by heat treatment as well as by alloying additions, as was shown for the S.A.E. 1050 steel pictured in Fig. 11. Whether the distribution of cementite is changed by heat treatment or by alloying additions will depend upon the relative costs, mechanical properties, size of specimen, etc.

The austenitizing characteristics of the steel are of particular importance in applications where a thin martensitic zone is desired (for example, a 0.020-inch case on a ½-inch bar). Extremely rapid heating is required to avoid excessive heat loss by conduction into the cold core, and this necessitates rapid austenitization at low temperatures. The furnace-cooled S.A.E. 2350 or 1350 steels or the sorbitic 1050 steel would be about equally satisfactory for such an application. The S.A.E. 4160 and 9255 may also meet the austenitizing requirements, although the results in Figs. 18, 19, and 20 would indicate otherwise. The furnace-cooled S.A.E. 1050 steel definitely would not meet the requirements because of the large quantity of free ferrite present.

The austenitizing TTT diagram (Figs. 5-12) and the hardness penetration curves (Figs. 15-20) show clearly the importance of heating to high temperatures (as compared to conventional furnace treatments) in order to obtain rapid austenitization, and thus enable steels to be induction-hardened in seconds instead of hours by furnace methods.

Hardening—Generally the induction-heated samples are quenched with a water spray from a quench ring which is a part of the induction hardening fixture. The austenite in the surface-heated zone will thus be transformed to martensite.

The role of alloying elements in hardening is to increase the hardenability or decrease the critical cooling velocity for the austenite to pearlite reaction. Generally the residual elements found in commercial steels will impart sufficient hardenability for most surface hardening applications; however, occasionally it is required to harden large sections through to the center and the deep hardening characteristics of the alloy steels will be needed. Also the shape of the part may be such as to prevent rapid quenching, and in that event, additional alloying elements may be desired to insure hardening in remote locations, such as the root of gear teeth.

The effect of the various alloying elements on the hardenability

has been determined accurately enough to permit direct calculations from the chemical composition (13), (14), (15). Manganese and nickel are useful elements for increasing the hardenability, and since they also improve the austenitizing characteristics they are to be favored over other elements, such as chromium, molybdenum, silicon, etc., for alloying additions to induction hardening steels.

The macrograph in Fig. 24 illustrates the "soft spots" which may result in induction-hardened steel when the quenching is not

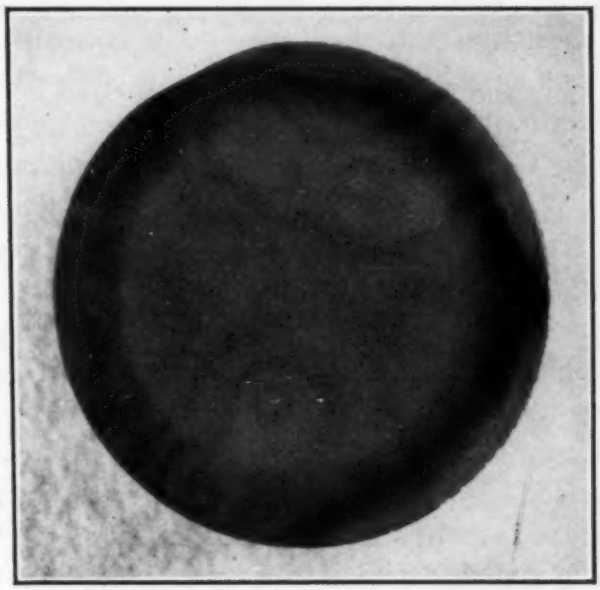

Fig. 24—Macrograph of Induction-Hardened L 1050 Steel Showing Soft Spots. × 2.5. The inner circle delineates the heated zone from the unaltered original core. The light areas in the heated zone are martensitic and the dark-etching regions consist of fine pearlite.

rapid enough to prevent the formation of pearlite. This particular steel was a laboratory prepared 1050 steel (0.47 C, 0.19 Mn, 0.13 Si) which had a critical cooling rate close to the quenching rate obtained in the hardening operation, and thus pearlite instead of martensite was formed in the heated zone. Increasing the manganese from 0.19 to 0.73 per cent in the commercial 1050 steel (Table I) was sufficient to increase the hardenability to the extent that soft spots did not occur in similar induction-hardened samples.

In view of the inhomogeneities which exist in the austenite, it may be desirable to specify a minimum hardenability for induction hardening steels in order to balance the effect of undissolved car-

bides and carbon concentration gradients in lowering the hardenability.

However, increasing the hardenability may have damaging effects since the cracking tendencies appear to increase with depth of hardening. The results of the five steels showed that cracking occurred only when the samples were deeply hardened. Thus, if induction heating is to be used for deep hardening of steel parts, special precautions (oil quenching, for example) may be necessary to prevent cracks forming during quenching.

Properties of Induction-Hardened Steels—The hardness of the martensite in induction-hardened samples is controlled primarily by the carbon content of the steel (8), (24), although the hardness noted in certain induction-hardened specimens is slightly higher (several points Rockwell "C") than that obtained by conventional methods of heat treatment. Several factors appear to be partly responsible for this difference in hardness. These are: 1—the high compressive stresses present in the martensitic zone of surface-hardened samples, and 2—a finer martensitic structure with less retained austenite which is frequently obtained by induction methods due to lack of homogenization during heating.

1. The effect of compressive stresses on the hardness is demonstrated in Fig. 23. Note that tempering at 100 degrees Cent. was sufficient to relieve the stresses and effect a drop in hardness for the shallow hardened sample without changing the hardness of the deeply hardened specimen.

Since compressive stresses in the martensitic layer also act to improve the fatigue properties by partially neutralizing external tensile stresses which cause fatigue failures, it is not always desirable or necessary to temper induction-hardened parts. However, there is no rigid rule and many samples of irregular shapes will require tempering to prevent cracking.

2. The effect of homogenization on the martensitic structure is indicated in Fig. 13. The sample heated 3 seconds is representative of induction-heated specimens. Note in Fig. 13a the absence of untransformed austenite in contrast to a large amount in the sample held for 3 hours. In view of these results it is to be expected that the induction samples with a fine martensitic structure will be slightly harder than furnace-treated specimens with coarse martensite and untransformed austenite, although it is well to remember that conversion of the soft austenite to martensite by a cold treatment

below freezing may be given to the furnace-treated sample to increase the hardness.

Summary

1. Data on the hardenability, austenitizing, and induction hardening characteristics of the following steels are presented: S.A.E. 1050, 1350, 2350, 4160, and NE 9255.

2. Hardenability was found to be of secondary importance in surface induction hardening of steel, although a minimum hardenability is required to prevent the formation of soft spots in the hardened case. Generally the manganese, silicon, and residual elements found in commercial plain carbon steels impart sufficient hardenability for most surface hardening applications; however, occasions may arise where the contour of the surface is such that rapid quenching is not obtained, in which event, deeper hardening, alloy steels may be required.

Hardenability was found to be of major importance for deep hardening by induction methods.

3. The rate of formation of austenite in the steels has been studied by following the hardness change, and the results have been presented as time-temperature iso-transformation (TTT) curves.

4. The austenitizing characteristics of a steel were observed to be closely related to the critical temperatures and the initial microstructure. The lower the critical values and the more uniform the distribution of cementite the more rapid was the formation of austenite at temperatures above the Ac_3.

5. The four alloy steels, in the furnace-cooled condition, were found to austenitize with greater ease than the S.A.E. 1050 steel (Fig. 10).

6. Altering the carbide distribution of the S.A.E. 1050 steel by a quench-temper heat treatment resulted in a great improvement in the austenitizing characteristics (Fig. 12). This steel with a sorbitic structure compared favorably with the furnace-cooled S.A.E. 1350 steel in its ability to transform to austenite.

7. The effect of increasing surface temperature has been studied by observing the changes in hardness and microstructure. The hardness results are presented as hardness penetration curves.

8. The ease of surface induction hardening a steel bar was found to depend primarily upon the austenitizing characteristics of

the material. The S.A.E. 1350 and 2350 steels which austenitized readily were observed to be best for surface induction hardening.

9. The effect of alloying elements on the induction hardening characteristics is related to their effect on the microstructure, critical temperatures, and hardenability. The ideal steel for induction hardening is one with low critical temperatures, little or no free ferrite, and medium to deep hardenability.

10. The slightly higher hardness of induction-hardened over furnace-hardened samples which has been frequently observed appears to be due to (a) compressive stresses in the martensitic zone of the surface-hardened samples, and (b) a finer martensitic structure with less retained austenite which is usually obtained by induction methods due to lack of homogenization during heating.

11. The results on the five steels showed that cracking occurred only when the samples were hardened to a great depth, and that the four alloy steels were more susceptible to cracking than the S.A.E. 1050 steel.

Acknowledgments

The authors are indebted to many persons for assisting in the experimental work and in the preparation of the paper.

We wish to express our appreciation to Messrs. W. E. Ruder, J. P. Jordan, J. E. Erb, and E. A. Smith for assistance and helpful discussions during the progress of the investigation and in the writing of the paper.

The members of the Industrial Heating and Welding Engineering Department, General Electric Company, in particular Messrs. Herbert Secor and R. A. Gehr, were most helpful in the induction hardening of the samples.

Our thanks also go to Mrs. C. B. Brodie and her staff for the photomicrographs.

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DISCUSSION

Written Discussion: By A. W. Herbenar, metallurgical engineer, Budd Induction Heating, Inc., Detroit.

The authors should be highly commended on their contribution of data constructive to induction hardening practice; especially, from the standpoint of the application of alloys of variant hardenability to such process.

The data presented in this paper show conclusively that the primary metallurgical factors involved in the induction hardening of steels are those associated with temperature and initial microstructure; and that the hardenability is only of secondary importance. However, certain predominating characteristics of alloying elements in regard to austenitizing rates are clearly shown.

The retarded rate of solution of ferrite in saturated austenite, as is characteristic of hypoeutectoid steels during rapid austenitizing, cannot be overemphasized. Examples have been brought to the attention of the writer wherein cracking during induction hardening, as well as fatigue failures on induction-hardened parts, could be traced to minute amounts of ferrite in the martensitic grain boundaries, a condition resulting from incomplete solution.

Although the limited solution time may be somewhat compensated for by heating to higher temperatures where the diffusion rates are accelerated, this may in certain cases prove detrimental from the standpoint of coarsening and added internal stresses.

Since a majority of applications of induction-hardened surfaces are in cases necessitating wear resistance, we have found that changing the composition in certain cases from hypoeutectoid to either eutectoid or hypereutectoid analysis is beneficial in two ways. First, it eliminates any possibility of incomplete ferrite

solution and, second, any residual carbides, either in spheroidal form or as "pearlite ghosts", contribute greatly to added abrasion resistance.

It also is of interest to note the retarded solution rate of a stabilized carbide phase as is exemplified by the S.A.E. 4160 analysis. Fig. 1 shows this steel to possess an extremely high hardenability when a homogeneous austenite is obtained by a comparatively long heating time at 850 degrees Cent. (1560 degrees Fahr.) prior to quenching. However, Fig. 18 shows that at 825 degrees Cent. (1515 degrees Fahr.) this steel shows a relatively shallow case; and it is only when temperatures of 900 to 975 degrees Cent. (1650 to 1790 degrees Fahr.) are reached that complete solution of the carbide phase occurs, allowing both the moderate carbide former (chromium) and the strong carbide former (molybdenum) to contribute a maximum effect on the hardenability. No such abrupt change is noticed in the alloys containing solid solution elements.

The authors clearly indicate that the "super-hardness" obtained in inductively hardened surfaces is at least partially due to the high compressive stresses in the martensitic zone; and further postulate that the fine martensite with less retained austenite frequently obtained by induction hardening is another contributing factor. Qualitative results obtained on internal hardening of cast iron substantiate their results. In plain gray irons, dimensional movement during cutting indicates very high compressive stresses in the hardened layer. In alloyed irons where retained austenite, which is relatively plastic, cushions compressive stresses accompanying the dilation to martensite; the compressive stresses are greatly reduced.

The present work which has correlated hardness gradients in relation to the hardenability of steels serves as a possible threshold for studies in stress distribution through the entire heated zone, including the martensitic region as well as the transition region.

AUTHORS' REPLY TO A. W. HERBENAR: We would like to thank Mr. Herbenar and the other gentlemen, whose discussions follow, for their interesting and valuable comments. We personally benefit from their remarks and criticisms, and feel that the readers of this paper will also benefit from their discussions.

We have little to add to Mr. Herbenar's remarks, except that we are in complete agreement with him regarding the damaging effect of small amounts of free ferrite in the hardened zone on the fatigue properties. The evidence at hand indicates that the fatigue cracks in gears and shafts are initiated in the ferrite when this phase is present in the hardened case.

Written Discussion: By James W. Poynter, metallurgist, Materials Laboratory, Air Technical Service Command, Dayton, Ohio.

Additional data on the metallurgy involved in the induction hardening of steel is always appreciated and especially so when, as in this case, a different approach to the problem is made. A number of things, however, detract from the value of this paper.

The errors inherent in the methods of temperature measurement and control used make the results obtained on the induction heating samples of value only in a general qualitative way and not quantitatively. While the currents induced in the thermocouple wires do not appreciably affect the reading of the instrument, the surface irregularities produced on the specimen by the spot welds attaching

the thermocouple are sufficient to cause localized overheating with resulting errors of appreciable magnitude in the observed temperature. The time lag in the conventional types of temperature measuring instruments, plus the delay in human response, introduces additional errors with the rapid heating rates employed.

Table III shows a maximum time of 6.9 seconds to heat a specimen to 750 degrees Cent. (1380 degrees Fahr.). This corresponds to a heating rate of more than 100 degrees Cent. per second. Even a total time delay of one-third second will result in a 35 degree Cent. error. It therefore seems entirely unjustified to show hardness penetration curves for two temperatures only 5 degrees apart (Fig. 17) or 20 degrees Cent. apart (Figs. 16 and 17). Even the 40 degree Cent. differences shown in Figs. 15 to 19 inclusive would appear to be of doubtful validity.

The reported hardness values are probably lower than the true values since, under certain conditions, cutting or heating the specimens (as would occur in mounting the sample in bakelite) produces a definite drop in hardness. Experimental evidence to substantiate this was given in the paper just presented at this session.

In the authors' discussion of the effect of internal stress the statement is first made that "as a result, high compressive stresses exit in the hardened layer", and later the seemingly conflicting statement that "the deeply hardened samples are likely to have the surface in tension". It will be appreciated if the authors will present an explanation as to why the direction of the surface stresses reverses in direction with increasing depth of hardening.

The validity of the explanation of the cause of the soft spots in Fig. 24 is questioned. The photograph of the induction heating fixture used (TRANSACTIONS, American Society for Metals, Vol. 34, page 357) shows that the turns of the inductor coil are supported by three spacers 120 degrees apart. Since these spacers are on the inside of the quenching ring, it appears likely that the soft spots shown, which are likewise 120 degrees apart, are due to improper quenching because of the shielding effect of the spacers. With the higher alloy steels, the transformation rate is sufficiently retarded so that the surface hardens even though the water spray is not uniform.

AUTHORS' REPLY TO JAMES W. POYNTER: Mr. Poynter has questioned our surface temperature measurements on the induction-heated samples, because of the possible effect of the following factors on the temperature measurements: (a) Overheating of the thermocouple wires, (b) the time lag of the temperature measuring device, and (c) the lag in human response. We recognized the importance of these factors, and have made a serious effort to minimize their effect on the temperature measurement.

To reduce any overheating of the thermocouple the individual wires were flattened, and spot welded a distance of approximately ⅛ inch apart. The temperature value measured is an average of the temperature of the area between the two wires of the thermocouple.

The temperature measuring device contained a string-type galvanometer which had a very rapid response to temperature changes. The heating rates for the various steels are recorded in Table III, and above the Curie temperature the

rates are less than 40 degrees Cent. (72 degrees Fahr.)—a value less than half that assumed by Mr. Poynter in his discussion. If we assume, as he did, a total time lag of one-third second the temperature correction factor will be less than 15 degrees Cent. (27 degrees Fahr.). In other words, our temperature values will be 15 degrees Cent. below the true temperature; but that does not mean that we are not justified in measuring temperature changes less than the correction factor of 15 degrees Cent. as Mr. Poynter indicates. We believe that the temperature differences shown in Figs. 15 to 19 inclusive are significant, and that the increase in depth of hardening with increasing temperature is proof that the temperature was higher for the samples as stated in the diagrams.

The best proof we can offer that our temperature values are only about 15 degrees Cent. lower than the true value is the heating curve shown in Fig. A.

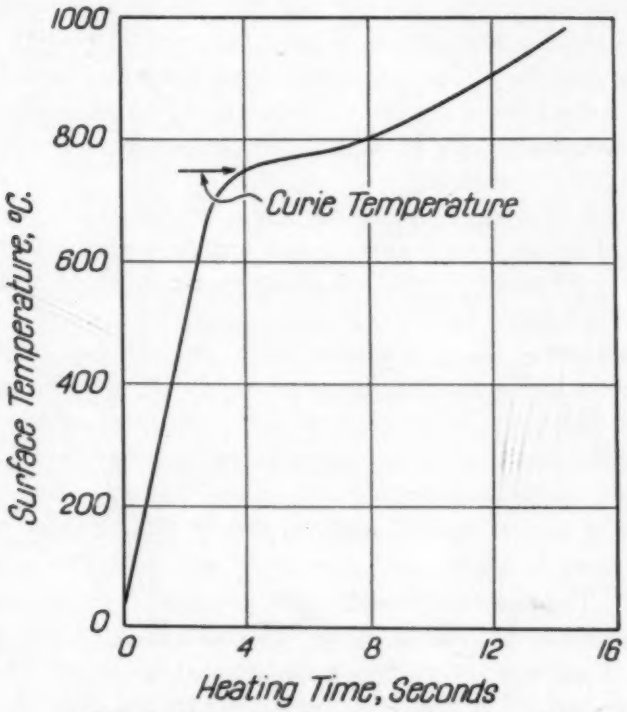

Fig. A—Induction Heating Curve for a Sample of Swedish Iron.

The Curie temperature is indicated by the change in heating rate at about 760 degrees Cent. (1400 degrees Fahr.), and compares favorably with a value of 760 to 780 degress Cent. by other methods of determination. We think that above the Curie temperature where the heating rates are slower, our temperature values are satisfactory for the purpose for which they were used.

We agree with Mr. Poynter that a drop in hardness may occur during cutting or mounting, and presented the information in Fig. 23 to show that the hardness dropped when the shallow induction sample was tempered at 100 degrees Cent.

He has asked us to explain the effect of depth of hardening on the type of residual stress pattern in the steel. Reference 31 covers that subject very thoroughly, but briefly the explanation is as follows: For shallow hardened samples the martensitic case is in compression, because it is not able to expand to its

natural dimensions. When the sample is hardened completely through, the surface metal is in tension (tangential stress), because the martensite in the center of the bar forms at a later time than the martensite at the outside. The center metal, therefore, is unable to expand to its natural dimensions, and the result is residual tensile stresses in the outer layer.

Mr. Poynter questions the validity of our explanation of the cause of the soft spots in Fig. 24, which was "that quenching was not rapid enough to prevent the formation of pearlite"; although, in his discussion he points out that the soft spots are due to improper quenching because of the shielding effect of the coil supports—another way of stating that the quenching was not rapid enough!

The fixture used for our induction hardening was satisfactory for hardening all the commercial steels. We included Fig. 24 to show that soft spots may occur if the hardenability of the steel is low; and that if the hardenability is low, care must be taken to see that the sample is quenched rapidly enough to prevent soft spots. Quenching may be of importance in some induction applications, and the solution is to alter the type of quench, or increase the hardenability of the steel.

Written Discussion: By H. B. Osborn, Jr., research director, Tocco Div., Ohio Crankshaft Co., Cleveland.

The authors of this paper are to be commended for the preparation of the data presented and which we are certain will enable present-day metallurgists to more thoroughly understand some of the problems facing them in making practical use of the induction heating process. In previous articles which we have had the occasion to present before the A.S.M. and other organizations, we have attempted to furnish control factors and other data to simplify the determination of suitable processing cycles. We have, however, admitted quite freely that such factors are only guides since the response of the material being treated is of the utmost importance and must be taken carefully into consideration.

For the sake of brevity and clarity, we should like to refer to the summary of the paper by Messrs. Martin and Van Note and comment as follows:

Paragraph 2: The presence of soft spots presumably in the surface area will in no way be affected by the hardenability characteristics of the material. Stress must be put on the use of a uniformly distributed quench which, if used, can provide a uniform rate of cooling well in excess of the critical cooling rate of the lowest hardenability steels. Please note that all high production surface hardening induction work is generally accomplished by the use of an inductor wherein the quench is supplied directly through its inside face. This precludes the possibility of soft spots being encountered when using a multiturn inductor necessitating the use of a quench which must be forced through the turns which enhances the formation of steam pockets in the vicinity of which low quenching rates are encountered. (See Fig. 1.)

It is possible to harden low hardenability steels to greater depths of hardness than can be obtained with conventional methods. This fact can be established by processing of large diameter rounds wherein the heated zone is cooled by both the quench water on the surface and the cold core beneath. Data on this were presented in TRANSACTIONS in an article by the writer listed as Reference 8 in the authors' bibliography.

Paragraph 8: We object to the statement that S.A.E. 1350 and 2350 steels

are best for surface hardening. This certainly should be qualified since inference from such statement would be that users of induction heating equipment should plan to use such steels. Obviously, the rapid austenitizing characteristics of these steels are of importance for extremely shallow hardening, but this type of application is very definitely in the minority. In ordinary production work adequate adjustment of processing cycles can be made with the use of the slow austenitizing steels by going to higher temperatures without obtaining any undesirable effects. Industry would certainly shy away from induction heating if it were unable to take advantage of the use of carbon steels.

In respect to Reference 9 the ideal steel for induction hardening may well be one with a low critical temperature, but from a practical standpoint we do not believe that this should be any criterion for the selection of the analysis. Due to the short heating times, we are able to use higher temperatures without coarsening and this is quite a common practice. For example, many production cycles are used on parts made of S.A.E. 1050 steel where to overcome the sluggishness of coarse initial structures with massive free ferrite, temperatures of 1700 to 1800 degrees Fahr. (925 to 980 degrees Cent.) are used. Obviously, a delay time follows such treatment in order that quenching may be accomplished at a lower temperature.

The induction hardening of crankshafts is perhaps the one application which involves the largest tonnage of material. With few exceptions, the steel used is an S.A.E. 1045 or 1050 with residual chromium. The presence of this small amount of alloying element narrows the free ferrite equilibrium range and makes it possible to use a modified normalizing treatment on the crankshaft forgings which produces a reasonably fine pearlitic structure with little, if any, massive ferrite. Such a structure may be readily austenitized.

Reference 10: In publications by the writer written in 1939, reference was made to the higher hardness obtained with induction hardening, and data were presented. A justification for this condition was offered on the basis of a more thorough carbide solution. Subsequent investigation, however, as detailed in additional publications by the writer, discounted any explanation on this basis and offered for consideration the possibility of the higher hardness resulting from compressive stresses and less retained austenite, in complete concurrence with the authors of this paper.

Reference 11: A general summary that the four alloy steels were more susceptible to cracking than S.A.E. 1050 is conceded as being true; however, regardless of the analysis, it is possible to modify the heating cycle to such an extent that cracking need not be a problem. Heavy sections of S.A.E. 52100 and WE-4150, as well as other similar high alloy analyses, have been through hardened as production items in sections up to 5 inches in diameter. The energy input per unit of surface area is maintained at low levels on the alloy steels so that the rate of heating is low and temperature differentials kept at a minimum by permitting a greater amount of heat flow by conduction. Furthermore, it has been our experience that cracking difficulties cannot always be blamed on the heating technique since quenching temperature, quenching rate, and final quenched temperature must be taken into consideration.

For example during the processing of 37 mm. A.P. projectiles made of S.A.E. 52100 steel, cracking was encountered when the steel was quenched at too high a

temperature and quenched too cold. An analysis of the cracking problem on this projectile seemed to indicate at first that the as-quenched temperature was entirely too low, but changes in the quenching cycle to permit removal of the projectiles at 200 degrees Fahr. did not solve the problem since at this temperature there is a considerable amount of retained austenite which converts after removal from the induction hardening equipment and evidences itself in the form of cracking. The solution to the problem was in using a long delay cycle which permitted the quench to take place from a temperature of several hundred degrees below the final austenitizing temperature, followed by a quenching cycle sufficient to bring the final temperature of the part down to approximately 100 to 150 degrees Fahr.

Induction heating does not accept, in principle at least, the need for oil quenching of any analysis regardless of hardenability. High alloy steels are generally heated at a slower rate than carbon steels and the quenching medium, whether it be oil, water, either submerged spray, direct spray, or tip type, depends entirely on the stresses which might be developed during the cycle. In other words, round cylindrical sections can be quenched with water under any circumstances, as can gears with large fillets and other parts not having sharp corners. Fine pitch gears, on the other hand, even made of carbon steel with low hardenability and low critical cooling rates, will generally require an oil quench.

REPLY TO DR. OSBORN: We agree with Dr. Osborn that hardenability is of secondary importance in surface induction hardening; however, it seems to us that his statement, "the soft spots . . . . . will in no way be affected by the hardenability characteristics of the material", is too broad. Certainly a *minimum* hardenability is required to prevent the formation of soft spots in any sample regardless of the type of quenching fixture. Fortunately, the commercial steels possess sufficient hardenability for almost every surface induction hardening application, and soft spots are not a general problem.

We agree that austenitizing characteristics, low critical temperatures, hardenability, carbon content, etc., are not the only factors to be considered in the selection of a steel for induction hardening. The physical properties, machinability, availability, and economic factors are equally important. The S.A.E. 2350 steel has excellent austenitizing response, but will have a limited application because of its high cost. The S.A.E. 1050 steel is a very popular steel for induction hardening, but it in turn is limited to applications where the heating rate is slow enough to allow solution of massive ferrite on heating.

Very little information is available on the subject of cracking in steel, and the comments by Dr. Osborn will be of practical value to all users of induction hardening equipment. If cracking occurs during quenching the solution of the problem may be to quench from a lower temperature, or to interrupt the quenching near the M_s temperature to allow equalization of the gradient (as in the martempering treatment), before cooling to room temperature.

Oral Discussion

W. WILSON, JR.:[6] The authors have characterized the results of isothermal austenite formation with two simple terms: half-hard and hard. Their results

[6]Associate metallurgist, Armour Research Foundation, Chicago.

should provide an insight into austenitizing treatments in general.

Quite often experimental work in the laboratory must be obtained by methods which lend themselves to precise control, e.g., isothermal studies of the formation of austenite from ferrite and cementite or the reverse process which is more frequently studied. Studies of the transformation of austenite under continuous cooling have shown a lag in the beginning of transformation when compared to isothermal results. The Grange and Kiefer method[7] is used when applying isothermal results to processes where temperature undergoes a continuous change. Would it not be quite likely that a similar situation exists in the formation of austenite?

Although grain growth in steels is better characterized by heats of steel than by chemical analysis, the discusser believes that some indication of the grain coarsening temperature-time relationship should be included as an upper limit. A paper by J. W. Poynter[8] recognizes that grain growth may occur during induction hardening.

With respect to the factor method for calculating hardenability, a paper appeared[9] showing the disagreement of various experimenters on these factors. This supports the hypothesis that the factors are dependent upon the other alloys in the steel. Therefore it would be practically impossible to calculate the hardenability of any steel from its chemical composition, as the authors stated in their paper.

REPLY TO DISCUSSION BY W. WILSON, JR.: Mr. Wilson has inquired about the relation between the formation of austenite during continuous heating and for isothermal conditions. We have not attempted a close comparison of our results for continuous heating with those obtained for constant temperature, but think that the hard and half-hard curves would be shifted to higher temperatures and longer times as the rate of heating increased in a manner similar to that found by Grange and Kiefer for the transformation of austenite on cooling.

Mr. Wilson's suggestion to include grain coarsening, temperature-time data in the austenitizing TTT diagrams has practical merit. Most of us have not given serious consideration to austenite grain size in induction hardening; although various investigators have shown that excessive coarsening may occur during induction heating. In our own work we have observed that excessive coarsening does not occur at temperatures as high as 1000 degrees Cent. (1830 degrees Fahr.) when the sample is heated extremely rapidly.

R. H. LAUDERDALE:[10] Martin and Van Note are to be commended for their fine work on the induction hardening characteristics of some of the more popular medium carbon steels. I am particularly interested in their study of the response to induction heating of furnace-cooled samples of S.A.E. 1050 steel versus quenched and tempered samples of the same steel having a sorbitic structure. The authors found that considerably longer heating time was required to fully harden the furnace-cooled samples than was required for the sorbitic structure.

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[7]R. A. Grange and J. M. Kiefer, "Transformation of Austenite on Continuous Cooling and Its Relation to Transformation at Constant Temperature," TRANSACTIONS, American Society for Metals, Vol. 29, 1941, p. 85.

[8]James W. Poynter, "Metallurgical Characteristics of Induction-Hardened Steel," TRANSACTIONS, American Society for Metals, Vol. 36, 1946, p. 165.

[9]*The Frontier*, March 1946, published by the Armour Research Foundation, Chicago.

[10]Metallurgist, Northern Pump Co., Minneapolis.

This effect of initial microstructure on the hardening time required is in perfect agreement with our findings at Northern Pump Company.

During the war we had the job of setting up procedures for surface hardening the teeth of thousands of gears used in the elevating and training mechanism of the 5-inch, 38-caliber anti-aircraft gun. These gears were made from S.A.E. 1045 hot-rolled bar stock or forgings. Two requirements had to be met by the heat treating department. First of all, the surface hardness of the teeth had to be between Rockwell C-50 and C-55. Secondly, only specified dimensional changes compensated for in machining could be tolerated. The gear teeth were induction-hardened after shaving and allowances were made for a shrinkage of 0.0015 inch on the pitch diameter and a runout of less than 0.0005 inch on the pitch diameter as a result of the hardening process.

I think you gentlemen will realize that these requirements were not easily met, especially on a water hardening steel. During the early days of the war when everything was being rushed to get into production, the gears in question were machined directly from bar stock or forgings with no special attention being given to the initial microstructure. As a result we had to change the hardening cycle for each batch of gears in order to meet the hardness requirements.

To illustrate, one batch of gears would require 4 seconds heating time at 140 kilowatts, whereas another lot of the same gears would require 7 seconds. Obviously the longer heating cycle caused too deep hardening and too much runout and shrinkage on the pitch diameter.

To enable production use of the shorter cycle which was desirable from all viewpoints, we soon adopted the procedure of producing a sorbitic structure on all gear blanks after rough machining. This was accomplished by water quenching from 1550 degrees Fahr. followed by drawing at 975 degrees Fahr., to give a hardness of around 225 Brinell.

It may be of interest to mention that in our present induction hardening work, which consists largely of producing thin cases of Rockwell C-64 to C-67 hardness on shuttle valves, valve plungers, and piston rods, we are using a 1 per cent carbon, plain carbon steel of tool steel quality, as we find it requires no initial heat treatment and responds more consistently to short hardening cycles than any other steel we have tried.

Our engineers feel that the premium price paid for this plain carbon tool steel is more than offset by the saving of the cost of the initial heat treatment, as well as in the guarantee that uniform hardness will be attained in repeat runs of the same job.

AUTHORS' REPLY TO R. H. LAUDERDALE'S DISCUSSION: Mr. Lauderdale's comments on the practical significance of initial microstructure on the induction hardening of gears, and his justification of the use of 1 per cent carbon steel for shuttle valves, valve plungers, and piston rods, are valuable information for engineers interested in applying induction hardening methods.

The use of the 1 per cent carbon steel is good evidence of the recent trend to the use of higher carbon steels for many induction hardening applications. As pointed out by Mr. Lauderdale, these higher carbon steels have a high metallurgical response which helps to insure uniformity in the hardened sample.

The austenitizing response of steel is improved by: (a) A heat treatment

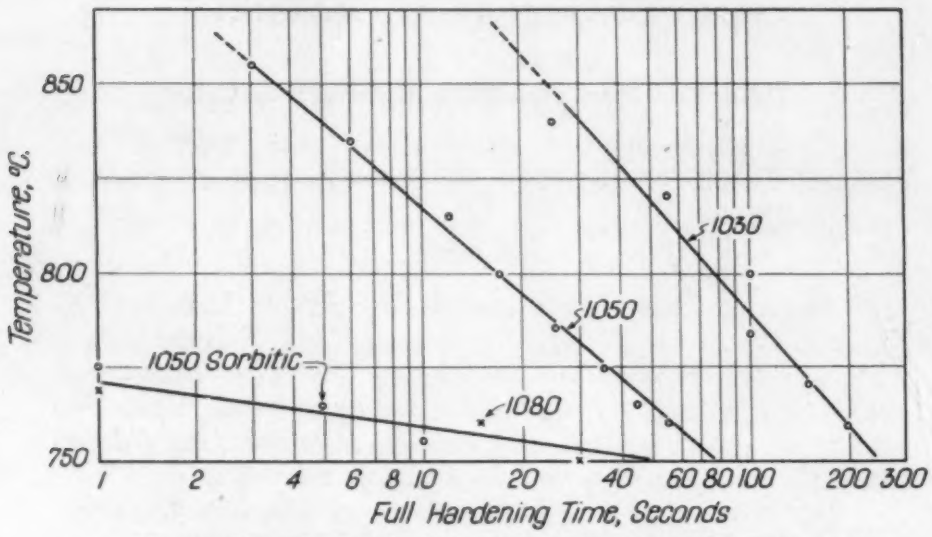

Fig. B—The Effect of Carbon and Heat Treatment Upon the Austenitizing TTT Curve.

or alloying additions which result in a more uniform distribution of the carbides; (b) increasing the carbon content; (c) lowering the Ac_1-Ac_3 transformation temperatures by additions of nickel or manganese. In Fig. B the effect of increasing the carbon content is compared with the effect of initial structure. Note that a sorbitic S.A.E. 1050 steel and a furnace-cooled S.A.E. 1080 steel have comparable austenitizing characteristics.

The method chosen to improve the autenitizing response of the steel will be governed by the physical properties, machinability, and economic factors. In the case cited by Mr. Lauderdale it was decided that the best material to use from an overall cost standpoint was a high carbon tool steel.

DEEP DRAWING ALUMINUM ALLOYS AT ELEVATED TEMPERATURES

Part I—Deep Drawing Cylindrical Cups

By Dan M. Finch, Scott P. Wilson and John E. Dorn

Abstract

The deep drawing properties of 3S-O, 52S-O, 24S-T, 24S-T86, 61S-T, XB75S-T and R301-T were evaluated by determining the maximum per cent draw that could be achieved by deep drawing cylindrical cups. The effects of punch radius, die radius, clearance, hold-down load, die temperature, and lubrication were studied.

Deep drawing aluminum alloys at elevated temperature results in a substantial and significant improvement in their drawability. In general, the drawability increases with increase in temperature. The maximum temperature for the deep drawing of the annealed alloys was limited to 700 degrees Fahr. (370 degrees Cent.) due to the fact that the lubricants which were studied were unsatisfactory above this temperature. Among the annealed alloys 52S-O exhibited the greatest improvement; the maximum height of cup at 700 degrees Fahr. (370 degrees Cent.) was slightly greater than twice the maximum height which could be drawn at atmospheric temperature.

The hardened alloys were studied at temperatures up to 450 degrees Fahr. (230 degrees Cent.). Higher temperatures were not used due to the possible effect of such temperatures on properties. Among the hardened alloys, XB75S-T, 61S-T and 24S-T86 exhibited the best drawability at 450 degrees Fahr. (230 degrees Cent.). These alloys have better drawabilities at 450 degrees Fahr. than the annealed alloys have at atmospheric temperature. Alloys 24S-T and R301-T at 450 degrees Fahr. (230 degrees Cent.) have drawabilities equal to the annealed alloys at 70 degrees Fahr. This fact suggests the possibility of using the high strength alloys for structural or semistructural applications of deep drawn parts by drawing at approximately 450 degrees Fahr. (230 degrees Cent.) rather than drawing the alloys in the annealed state at atmospheric temperatures and heat treating after forming.

A paper presented before the Twenty-seventh Annual Convention of the Society held in Cleveland, February 4 to 8, 1946. Of the authors, Dan M. Finch is research engineer, Scott P. Wilson is engineer, and John E. Dorn is project supervisor and associate professor, University of California, Berkeley, Cal. Manuscript received March 12, 1945.

Forming by Deep Drawing

THE forming of sheet material into shells, vessels, boxes, and other articles possessing considerable depth is usually best accomplished through the use of the deep draw press provided the number of parts to be produced is sufficiently great to warrant the high die costs.

The process of deep drawing is most readily described in terms of the simplest deep drawing operation, namely, the drawing of a

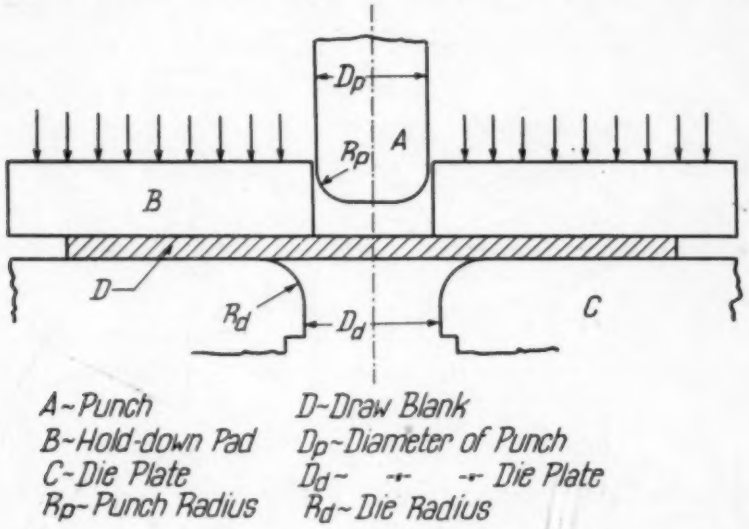

Fig. 1—Essential Parts of a Deep Draw Press.

cylindrical cup from a circular blank. Due to the symmetry of the product both during and after forming, cylindrical shells lend themselves most readily to theoretical and practical considerations.

The deep draw press consists essentially of three component parts, namely, the punch, the hold-down pad, and the die plate. A schematic diagram of a deep draw press, together with a circular blank to be drawn, is shown in Fig. 1. During drawing the punch (A) descends until it contacts the blank to be drawn (D), at which time further motion of the punch results in a radial displacement of all circumferential elements in the flange towards the punch. The hold-down pad (B) serves to prevent wrinkling of the flange during drawing. Thin sheets usually require hold-down forces to prevent wrinkling; as a result of their greater resistance to buckling, thick sheets may often be successfully drawn without the use of hold-down forces.

The stresses imposed upon a circular blank as it is drawn into a

cylindrical cup can be visualized by reference to Fig. 2 which shows a wedge-shaped section of a partially drawn cylindrical cup. During the first stages of drawing, the blank is bent around the radius of the punch. This bending action begins with the start of the draw and is practically complete, for cases where the clearance between the punch and die is small, when the punch stroke is equal to the sum of the punch and die radii. As drawing continues beyond this position the wall section of the cup is subjected to simple tensile stresses. As various elements of the blank are drawn around the die radius they

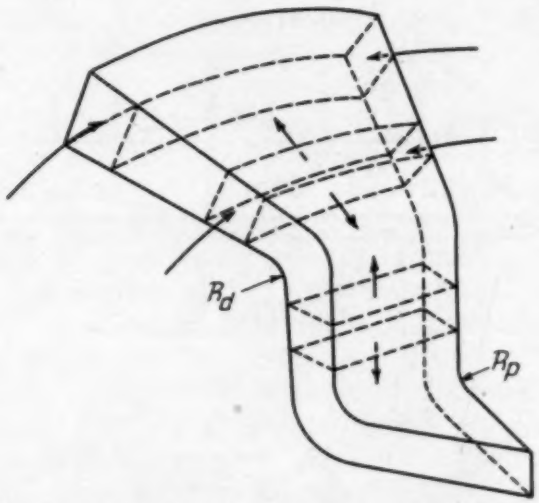

Fig. 2—Wedge of a Cylindrical Cup Partially Drawn.

bend to conform to this radius and when they become tangent to the wall of the cup they unbend. Throughout the process of drawing the elements in the wedge portion of the flange shown in Fig. 2 are displaced radially toward the punch.

Secondary complicating factors, however, serve to modify the above simplified explanation.

1. As drawing takes place the section under the hold-down thickens. The greatest thickening occurs at the outer fiber of the flange where principally simple circumferential compression stresses exist. Therefore the thickness of the flange increases during the drawing operation.

2. The tensile stresses introduced in the side wall of the cup cause the cup wall to thin, the thinning usually being greatest immediately above the radius of the bend at the bottom of the cup.

3. Friction between the hold-down plate and the blank results in an increase in the radial tensile stresses in the flange. After a small

initial displacement the hold-down is in contact with only the outer fiber of the flange since the thickening is greatest at this fiber.

4. The hold-down also introduces a compression stress in the thickness direction of the flange at its outer fiber. In general, however, the stresses so introduced are so small with respect to the radial or circumferential stresses that their influence on the final dimensions of the cup may be neglected.

5. If the clearance between the punch and the die is insufficient to accommodate the thickened wall of the cup, ironing takes place. Under these conditions the wall of the cup is squeezed between the punch and the wall of the die to a thickness equal to the available space between the punch and the die.

The processes occurring during the deep drawing of metals have been described in numerous articles in the literature. Asimow and Crombie (1)[1] list a bibliography of 248 references. Although no attempt will be made here to review the voluminous literature on this subject, the reader is referred to Sellin (2) "Handbuch der Ziehtechnik" and Jevons (3) "The Metallurgy of Deep Drawing and Pressing" for details not completely covered in this report which is based on work done under Project NRC-548 for the Office of Production Research and Development of the WPB under the supervision of the War Metallurgy Committee and with the co-operation of the aircraft industry. Sachs' (4) "Spanlose Formung" also deserves special attention inasmuch as it presents the most satisfactory analysis of the cup drawing operation currently available.

The theory on the conditions of wrinkling and the accurate solution of the conditions of bending and unbending around the die radius have not been satisfactorily introduced into the analysis of the deep drawing problems. Until these advances are made the evaluation of the drawability of metals must remain experimental.

Cupping Tests—Inasmuch as metal undergoing a deep drawing operation is subjected to nonuniform action of combined stresses, the simple tension test and the bend test do not provide sufficiently complete data to permit an evaluation of drawability. Various cupping tests have been devised in attempts to establish a simple standard procedure for rating the drawability of metals. In cupping tests, such as the Erichsen, Olsen, Amsler and Avery Tests (3), a round-nosed punch or a ball is forced into sheet which is held over a die ring. The maximum depth of the cup which can be produced with-

[1]The figures appearing in parentheses pertain to the references appended to this paper.

out tearing is used as a criterion of the relative cupping values. The Guillery Test resembles the above mentioned cupping tests except that the test is stopped when the pressure on the punch reaches its maximum value; the depth of the cup so produced gives the cupping value. Other tests such as the Jovignot, N.P.L. and K.W.I. tests are described by Jevons (3). Many of the cupping tests approximate a bulging operation more closely than a deep drawing operation and consequently they do not correlate accurately the deep drawing characteristics of metals. Sachs (5) suggested a wedge draw test which approximates the conditions of stress in the flange of a cup as it is being drawn. Unfortunately this test introduces undesirable frictional resistance at the sides of the wedge and it does not introduce the bending and unbending of the flange which occurs in the deep drawing operation. Consequently the effects of punch and die radii are not evaluated by this test. It is now generally agreed that a deep drawing test should reproduce as accurately as possible all conditions encountered in actual production operations (6). The simplest deep drawing operation consists of the production of the cylindrical cup from a circular blank. This procedure was selected for the evaluation of the deep drawing of aluminum alloys reported herein. Dahle and Boulger (7) have demonstrated the utility of this method by correlating the drawability of steels using the values obtained from draw tests on cylindrical cups.

Experimental Equipment and Technique

The Draw Press—A subpress unit consisting of a heated drawing die, heated or water-cooled punch, and a heated pneumatic hold-down pad was used to draw cylindrical cups from circular blanks. The subpress was installed between the upper and the lower platens of a variable speed hydropress. Reference to the photograph of Fig. 3 and to the schematic drawing of the installation given in Fig. 4 will show the general assembly and construction of the subpress.

The main frame of the subpress was constructed of welded channel iron. A 4½-inch diameter water-cooled ram was operated through a guide bushing in the top cross member. A 3-inch diameter punch was attached to the ram and carefully aligned with the drawing die. Die insert rings were provided in order to vary the draw radius and clearance between the punch and die. Provision was also made to change the punch so that the punch radius could be varied.

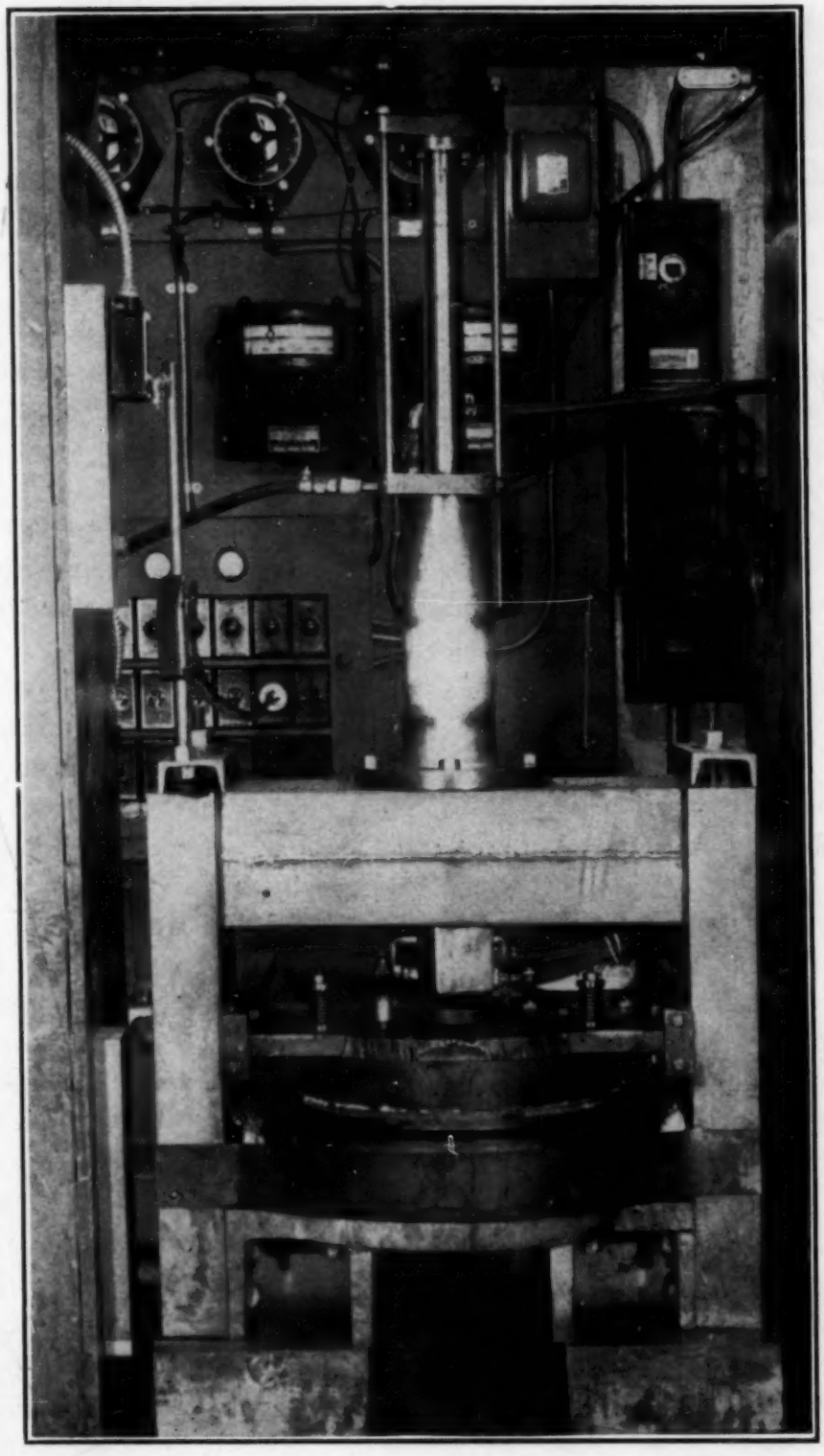

Fig. 3—Draw Press Assembly.

The hold-down mechanism consisted of an air operated piston and cylinder in the shape of an annular ring with a central hole large enough to allow the punch and ram to pass through. A pressure regulator and calibrated gage were placed in the line to maintain constant any selected hold-down load during the complete drawing operation. The weight of the moving parts was counterbalanced by a set of three springs. The mechanism was calibrated for conditions identical to those used for drawing.

The surfaces of the die plate and hold-down plate were ground and polished. The die ring inserts and punches were case hardened and polished. The radii were machined using a special lathe jig to provide a cross section having a true circular drawing surface tangent to the die plate and die bore.

Electric heaters were used in the die plate and hold-down pad at locations 11 shown in Fig. 4. Four annular ring heaters totaling 6050 watts were installed in the die plate. Two annular ring heaters totaling 2250 watts were used in the hold-down pad.

Temperature Control—The temperature of the die plate and hold-down pad was maintained at the desired setting by means of on-off controlling pyrometers. The mass of the metal parts smoothed out this periodic input to give a cyclic variation of $\pm$ 6 degrees Fahr.

Temperature traverses for each setting were made across the die plate with a 13-inch diameter, 0.102-inch thick dummy specimen in which five thermocouples were placed at various distances from the center of the blank. Typical surveys are plotted in Fig. 5 and show a maximum deviation of $\pm$ 8 degrees Fahr. from the nominal temperature setting. The curves showing effect of temperature on drawability indicate these variations did not materially influence the per cent draw.

Preheat Time—A number of preliminary tests were made to determine the minimum time required to preheat a specimen before drawing. The minimum preheat time is the interval necessary between the time the room temperature blanks are placed on the heated die pad and under the hold-down pad until the specimen has reached the desired temperature. To determine this preheat time a dummy specimen 0.064 inch thick and 7.5 inches in diameter with a thermocouple embedded 2 inches from the center was placed on the die plate and held down similar to a specimen in an actual test. Readings were taken at intervals to determine the temperature-time curves shown in Fig. 6. A rapid increase in temperature was obtained dur-

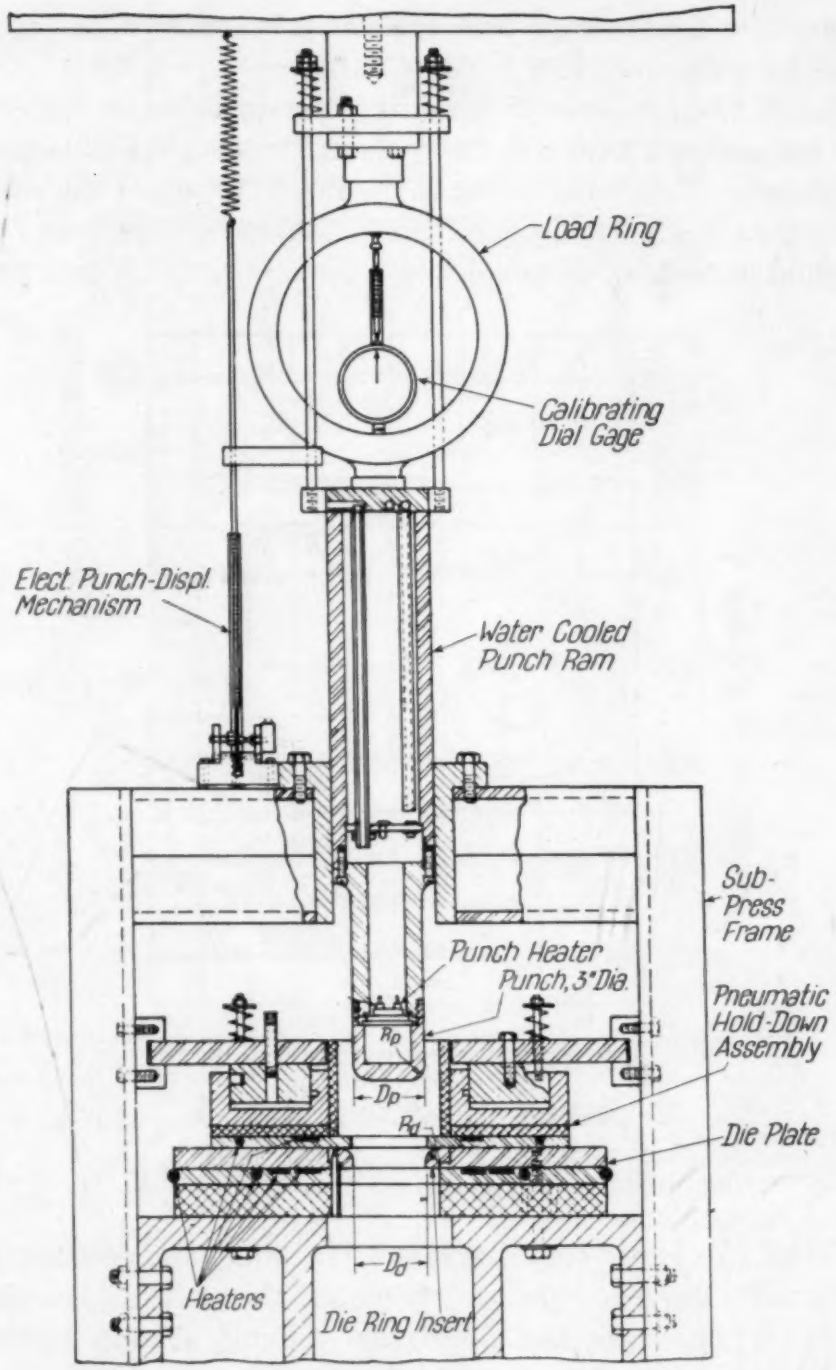

Fig. 4—Sub-Press Assembly Deep Draw Press.

ing the first minute but in all cases the temperature was constant after 5 minutes. On the basis of these tests, 5 minutes was adopted as the standard preheat time. For a series of tests at 700 degrees Fahr. (370 degrees Cent.) it was necessary to reduce the preheat time to 2 minutes in order to prevent excessive charring of the lubricant.

The standard 5-minute preheat time may be reduced in production practice by preheating in a separate furnace.

Punch Temperature—Preliminary investigations on the effect of punch temperature revealed that greater draws were obtained with a cold punch. This result is due to the rapid cooling of the side walls of the cup as it is being drawn, thereby increasing their load carrying ability and permitting deeper draws. Consequently all data reported

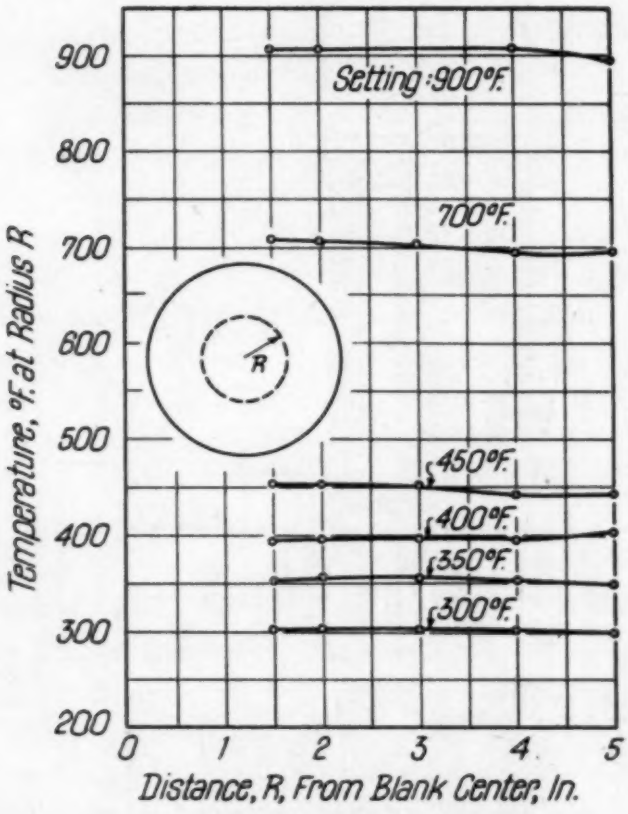

Fig. 5—Temperature Traverse of a Blank Under the Hold-Down Pad for Various Temperature Settings.

herein refer to a water-cooled punch. The punch temperature varied from 70 to 80 degrees Fahr. depending on the ambient temperature.

Blank Preparation and Centering—Circular blanks 0.064 inch thick were used for all tests. The blank sizes were varied in steps corresponding to 2.5 per cent draw as shown in Table I. The blanks for the tests on the annealed alloys were prepared by sawing the sheet to the proper size using a circle cutting attachment on a metal cutting band saw and by deburring the rough edge. The blanks used for tests on the hardened alloys were sawed oversize and turned and polished on a lathe. A small number of sawed-edge specimens of the

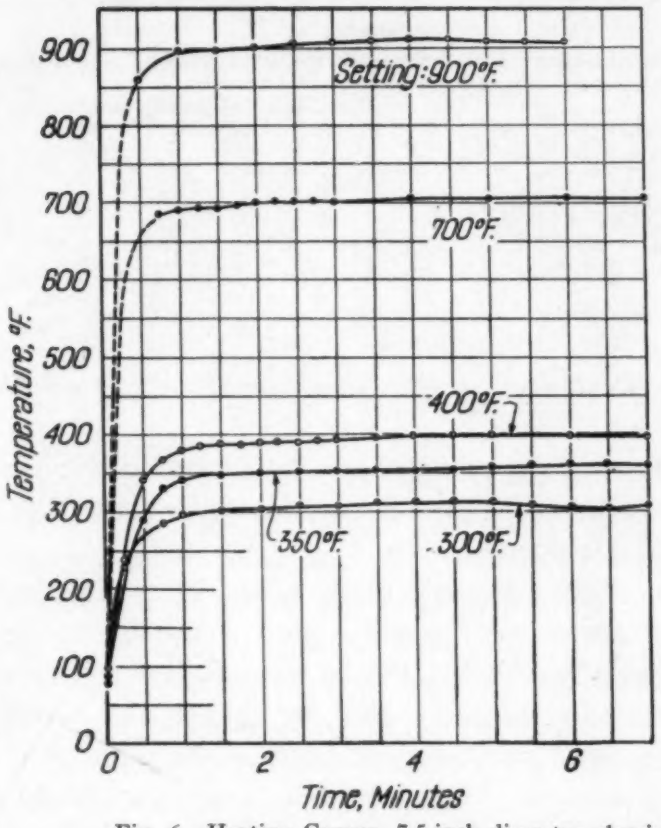

Fig. 6—Heating Curves. 7.5-inch diameter aluminum specimen, 0.064 inch thick. Thermocouple located 2 inches from center of specimen.

Table I

Draw Blank Sizes

(Each specimen was stamped with the direction of rolling, alloy, sheet number and diameter. All blanks were 0.064 inch thick.)

$$\text{Per Cent Draw} = \frac{D_b - D_p}{D_b} \times 100$$

$$\text{Drawability} = \frac{D_b}{D_p}$$

D_b = Blank Diameter
D_p = Punch Diameter

Per Cent Draw	Drawability	Blank Diameter (D_b) Inches
25	1.33	4.00
27½	1.38	4.14
30	1.43	4.29
32½	1.48	4.44
35	1.51	4.62
37½	1.60	4.80
40	1.67	5.00
42½	1.74	5.22
45	1.82	5.45
47½	1.91	5.72
50	2.00	6.00
52½	2.10	6.31
55	2.22	6.66
57½	2.35	7.06
60	2.50	7.50
62½	2.66	8.00
65	2.86	8.57
67½	3.08	9.23
70	3.33	10.00

Table II
Alloys and Temperatures Investigated

Group	Alloy	Die Temperatures—Degrees Fahr. 70	300	350	400	450	700
1A	3S-O, 52S-O	X			X	X	X
1B	24S-O	X			X	X	X
2A	24S-T	X	X	X	X	X	
2B	61S-T, 24S-T86	X	X	X	X	X	
	R301-T, XB75S-T	X	X	X	X	X	

Using Punch and Die Radii = 8t (0.512 inch), Punch Temp. = 70-80 degrees Fahr. (water-cooled).

hardened alloys were used in spot checking the effect of edge condition.

The specimens were centered on the die plate with a centering head fixture provided with graduated adjustable arms that could accommodate all blank diameters.

Lubrication—Several lubricants were used in the present investigation. All lubricants were applied with a brush leaving a heavy coating on both sides of the blank. In some earlier tests with 24S-T, a hot forging agent was used. For the other tests a 600W steam cylinder oil mixed with powdered graphite in the ratio of 1¼ pounds graphite per gallon of oil. The lubricants investigated and the lubrication problems encountered are discussed more fully in the section on Results and Discussion.

Punch Speed—The majority of the tests were carried out with an average speed of about 26 inches per minute. A few runs were made on several of the precipitation hardened alloys at 450 over a range of speeds from 10 inches per minute to 145 inches per minute.

Test Program—The maximum per cent draw and the maximum height of cup using the minimum hold-down load just sufficient to prevent wrinkling have been determined for the alloys and temperature conditions given in Table II.

Materials Investigated

The aluminum alloys that were tested for their deep drawing properties are listed in Table II. The physical properties of the sheets used and the corresponding typical tensile properties are shown in Table III.

The grain counts reported were prepared in accordance with A.S.T.M. Spec.. E2-39T for nonequiaxed grains. The counting plane was longitudinal and perpendicular to the rolled surface. Line a-a

Table III
Actual and Typical Physical Properties of 0.064 Aluminum Sheet

Alloy	Grain Size* Grains Per Mm. Line a-a	Line b-b	Tensile Yield Strength Set 0.2 Psi. Actual*	Typical	Ultimate Strength Psi. Actual*	Typical	Elongation in 2 Inches Actual*	Typical
3S-O	....	....	6,750	6,000	16,500	16,000	31.3	30.0
52S-O	....	....	12,150	14,000	27,300	29,000	26.1	25.0
24S-O	....	....	13,550	10,000	32,700	26,000	17.5	20.0
24S-T	5.8	21.4	48,300	46,000	70,300	68,000	19.6	19.0
24S-T86	12.8	47.8	71,250	73,000	75,650	77,000	5.6	6.0
XB75S-T	9.4	36.5	72,000	.....	83,500	.....	12.3	....
61S-T	21.5	64.7	39,500	39,000	45,950	45,000	12.2	12.0
R301-T	20.6	45.9	60,100	60,000	67,200	68,200	9.9	10.0

*Average values of sheets used.

was parallel to the direction of rolling. Line b-b was perpendicular to line a-a.

The alloys which were investigated are classified into four groups as shown in Table II in accordance with the general effects of exposure to elevated temperatures on their atmospheric properties. This method of classification permits tentative suggestions relative to the maximum permissible drawing temperatures for each class of alloys. The various groups have the following characteristics:

Group 1A. Annealed alloys which are not precipitation-hardenable.
Alloys in this group may be deep drawn at any temperature up to their hot short or grain growth temperature without introducing significant changes in properties.

Group 1B. Annealed alloys which are precipitation-hardenable at some temperature.
Alloys in this group may be deep drawn at any temperature up to their hot short or grain growth temperatures. In order to obtain full advantage of the potential properties of these alloys they are generally solution heat treated, quenched and aged after forming.

Group 2A. Alloys which are solution heat treated and precipitation hardened at atmospheric temperature.
Group 2A alloys may yield precipitation hardening effects at temperatures up to about 450 degrees Fahr. (230 degrees Cent.) dependent upon the alloy and time at temperature. For longer times at temperature or for higher temperatures overaging effects may be encountered. Consequently the properties of Group 2A alloys may be improved by holding at the forming temperature for certain ranges of time and temperature. For other con-

ditions such as higher forming temperatures or longer times at temperature, softening may result.

Group 2B. Alloys which are solution heat treated and precipitation hardened at elevated temperatures.

Group 2B alloys which have been precipitation hardened at elevated temperatures may yield only minor amounts of additional precipitation hardening effects in the lower range temperatures. Softening resulting from overaging may occur at higher temperatures or for longer holding times at the lower temperatures.

Group 2A and Group 2B alloys, therefore, should not be drawn at temperatures which yield permanent softening. The exact temperatures and times that yield appreciable softening will be reported at a later date. The data on the effect of time and temperature on aging suggest that 450 degrees Fahr. (230 degrees Cent.) is about the maximum value for the higher temperature aging alloys of Groups 2A and 2B. The lower temperature aging alloys may exhibit reductions in their atmospheric temperature properties when held at lower temperatures for relatively short times. Small variations in timing the exposure of Group 2A and Group 2B alloys to temperatures above 450 degrees Fahr. (230 degrees Cent.) will likely result in wide variations in properties. The suggested temperature of 450 degrees Fahr. (230 degrees Cent.), however, was made tentatively in order to define the temperature ranges for the present investigation on deep drawing so that the scope of this investigation could be kept within reasonable limits. Inasmuch as the selection of 450 degrees Fahr. (230 degrees Cent.) as the maximum temperature for investigations on the drawability of Group 2A and Group 2B alloys was based exclusively on the physical effects of temperature, additional investigations on corrosion, stress corrosion, endurance strengths, and other properties will be required for those temperatures which appear to be promising forming temperatures.

Results and Discussion

Evaluation of Drawability—The significant measures of the drawability of metals are those that reveal the maximum successful draw that may be achieved in a single operation. Although standard measures of drawability have not yet been established, three measures which have been used have sufficient virtue to warrant consideration. According to the first, the per cent draw is defined by:

$$\text{P. D.} = \frac{(D_b - D_p)}{D_b} \times 100$$

where P. D. = per cent draw
D_b = blank diameter
D_p = punch diameter

The second measure of the maximum permissible draw is called the drawability. According to this measure the drawability is defined by the following dimensionless ratio:

$$D = \frac{D_b}{D_p}$$

where D = drawability
D_b = blank diameter
D_p = punch diameter

The third measure is based on the dimensionless ratio of the final height of the cup to one-half of the cup diameter according to the equation:

$$\text{R. H.} = \frac{h}{1/2\, D_p}$$

where R. H. = relative height
h = final outside height of cup
D_p = diameter of the punch

The above definitions appear adequate for cases where the sheet thickness is small relative to the diameter of the punch.

The various measures of drawability are not exactly comparable inasmuch as the per cent draw and the drawability are measures of the initial blank and punch diameters whereas the relative height of the cup refers to a drawn cup in which the walls may have thinned in certain regions and thickened in others. It is important to realize that for a constant per cent draw, the relative height of a drawn cup may be altered by changing the hold-down load, clearance or any other factor which causes a change in the drawing force and thereby a variation in wall thickness.

Types of Failures—Several types of failures were noted during the deep drawing of aluminum alloy sheet as shown in Fig. 7. Failures A to F limit the maximum draw that may be achieved under the conditions of test which were investigated. G and H were avoidable by increasing the hold-down force. The hold-down force was regu-

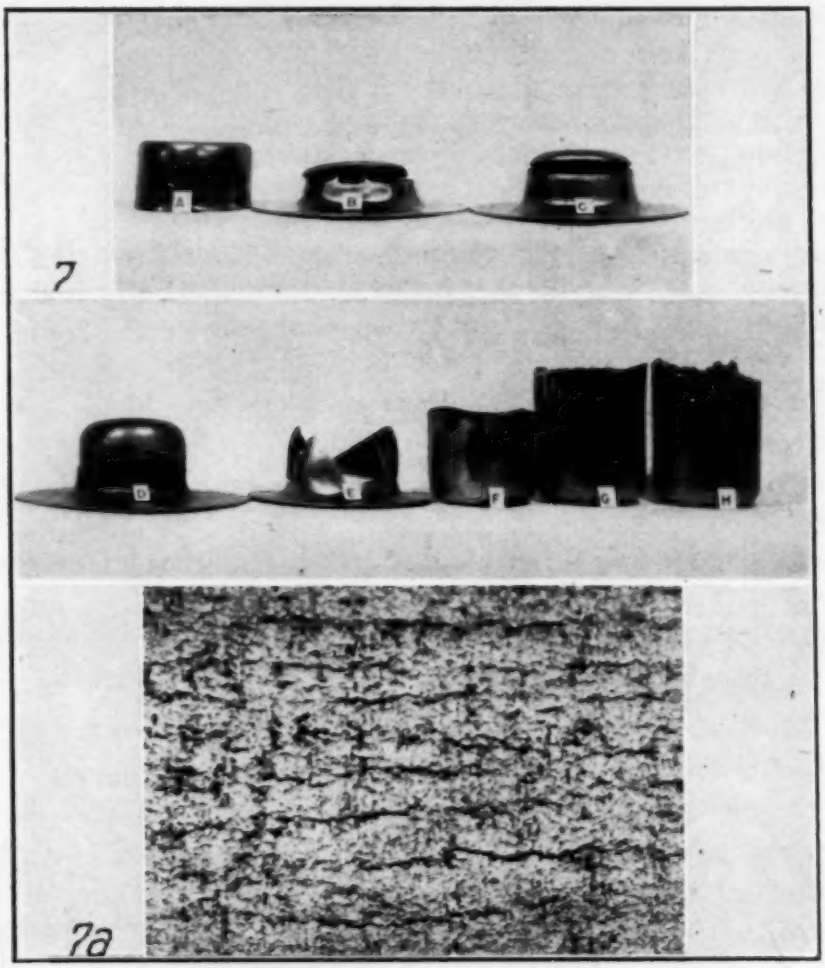

Fig. 7—Types of Failure for Deep Drawn Cups. A—Bend cracks. B—Bend cracks leading to rupture. C—Radius tear. D—Side wall tear. E—Dome rupture. F—Lip failure. G—Wrinkles. H—Wrinkling with ironing tear.
Fig. 7a—Bend Cracks. Macrograph—× 10.

lated to yield no more than the slight wrinkling shown in Fig. 8.

Lubrication—One of the major difficulties encountered in this investigation centered around suitable lubricants for deep drawing at elevated temperatures. Commercial lubricants and special lubricants compounded by several oil company research organizations were tried at temperatures up to 800 degrees Fahr. (425 degrees Cent.). A petroleum base hot forging agent was satisfactory up to 450 degrees Fahr. (230 degrees Cent.) but at this temperature smoking occurred. At higher temperatures the residue that remained was very gummy and charred on the specimen and die. A 600W steam cylinder oil was usable for some alloys at 700 degrees Fahr. (370 degrees Cent.) but did not eliminate metal pick-up on others. It was found that the

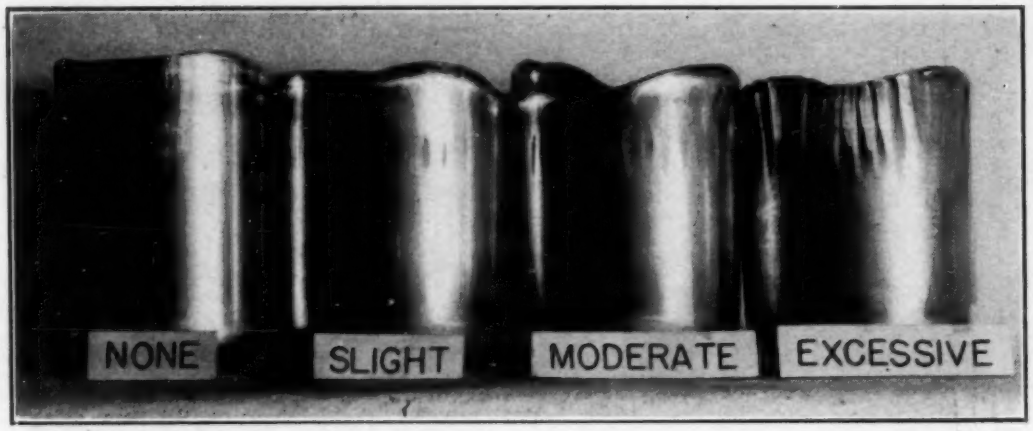

Fig. 8—Wrinkle Standards. Acceptable: None to slight (wrinkles less than 0.005 inch deep). Not acceptable: Moderate to excessive (wrinkles more than 0.005 inch deep).

addition of No. 620 graphite in the ratio of 1¼ pounds of graphite per gallon of oil would eliminate metal pick-up for all of the alloys investigated for temperatures up to 700 degrees Fahr. (370 degrees Cent.). At this temperature the oil residue charred and baked on the metal parts exposed to air if held at temperature for too long a time. Excessive charring was avoided by reducing the preheat time to 2 minutes and wiping the subpress parts clean after each draw.

Although the 600W steam cylinder lubricant modified with graphite prevented metal pick-up up to 700 degrees Fahr. (370 degrees Cent.) and was used as the lubricant for the majority of the tests for this investigation, it is likely that better lubricants for elevated temperature deep drawing of these alloys may be formulated.

Die Design: Surface Finish and Hardness—The surface finish and hardness of the die were found to have an important influence on deep drawing of aluminum alloys at elevated temperatures. A smooth polished surface is necessary to prevent metal pick-up and a hard surface is necessary to prevent scoring of the die radius. After a preliminary survey on the effect of the die condition of deep drawing at elevated temperatures, all dies were case hardened and polished.

Edge Condition of Blank—The finish and condition of the edge of the blank had little effect upon the drawability of the annealed alloys, 3S-O, 52S-O and 24S-O of Groups 1A and 1B. Specimens with sawed edges were used for most of the tests upon these alloys.

The preliminary tests upon sawed blanks of 24S-T alloy revealed that most failures were due to lip fractures. All specimens of this alloy were subsequently cut oversize and lathe turned and pol-

Table IV
Earing

		70° F.			450° F.		
Group	Alloy	Number of Ears	Degree	Orientation to Direction of Rolling	Number of Ears	Degree	Orientation to Direction of Rolling
1A	3S-O	.	Very Slight	...	2	Slight	Random
	52S-O	4	Slight	0° & 90°	4	Large	0° & 90°
1B	24S-O	4	Moderate	45°	4	Large	45°
2A	24S-T	4	Very Slight	45°	4	Moderate	45°
	24S-T86	4	Slight	45°	8	Slight	22.5°
	61S-T	8	Very Slight	...	8	Slight	22.5°
2B	R301-T	4	Slight	45°	4	Slight	45°
	XB75S-T	.	None	...	4	Slight	45°

ished to size. This procedure permitted greater draws for the cases yielding lip fractures. Check tests on blanks with sawed edges indicated the hard alloys other than 24S-T were not particularly sensitive to edge condition. All the hardened alloys were turned and polished to eliminate the possible influence of edge conditions on the results.

Earing—Pronounced ears were observed on deep cups drawn from 24S-O and 52S-O in the temperature range 400 to 850 degrees Fahr. (205 to 455 degrees Cent.). All of the alloys showed a tendency to form ears at the higher temperatures and a majority produced ears to a smaller degree at 70 degrees Fahr. The number, orientation, and magnitude of the ears varied according to the alloy and temperature as shown in Table IV. These ears may also be seen in the photographs in Figs. 9 and 10.

Maximum Permissible Drawing Temperature—In the present investigation, lubrication was found to be the controlling factor in establishing the maximum permissible forming temperature for Groups 1A and 1B alloys. Although an exhaustive search was not made, all lubricants which were tried were found to be unsuitable for deep drawing these aluminum alloys at temperatures above 700 degrees Fahr. (370 degrees Cent.). The factors involved were (a) the breakdown of the lubricating film yielding metal pick-up on the die and (b) gumming and charring of the lubricant which required frequent cleaning of the die parts.

Microstructure studies of the Group 1A alloys revealed that no significant change in grain size occurred at temperatures up to 900 degrees Fahr. (480 degrees Cent.). Group 1B alloy 24S-O showed no significant increase in grain size at temperatures up to about 850 degrees Fahr. (455 degrees Cent.) where this material was hot-

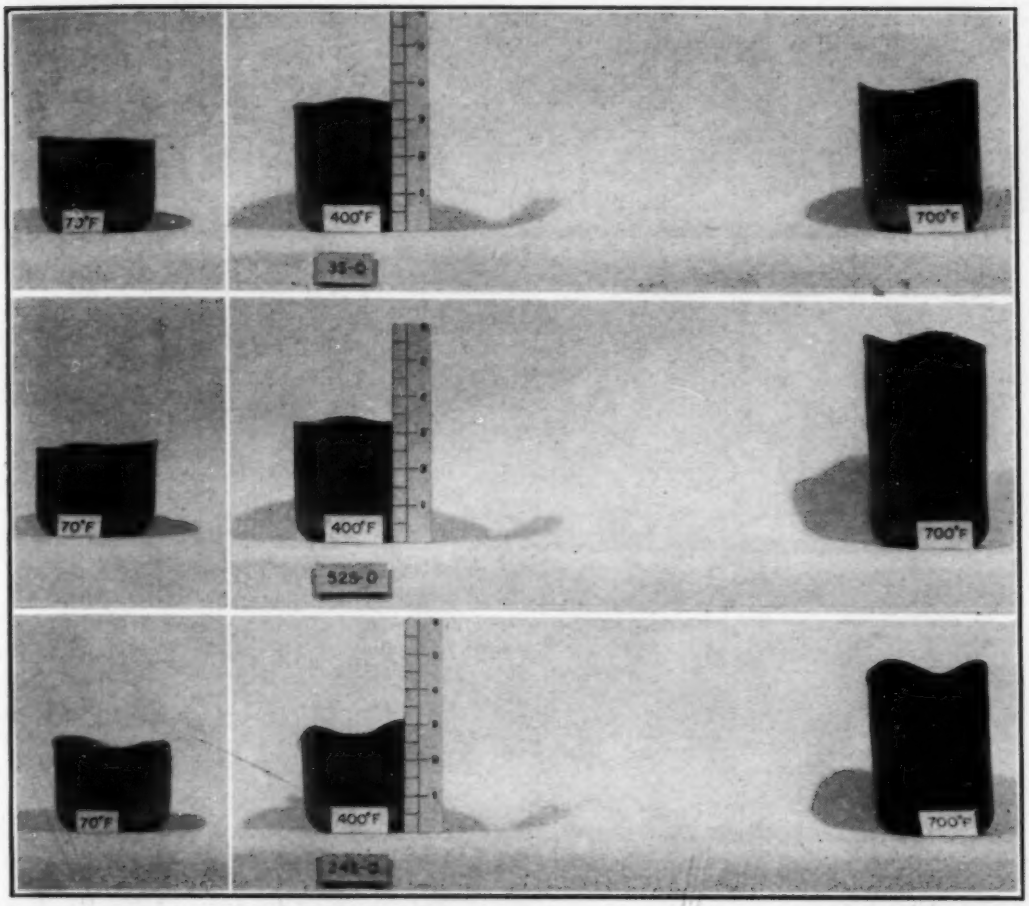

Fig. 9—Maximum Successful Draw at Various Temperatures for Annealed Temper Aluminum Alloys.

short. Until a suitable lubricant for use at temperatures above 700 degrees Fahr. (370 degrees Cent.) can be found, however, the maximum drawing temperature for the Groups 1A and 1B alloys is limited to approximately 700 degrees Fahr. (370 degrees Cent.).

For the Groups 2A and 2B alloys the maximum temperature is limited by the metallurgical effects of elevated temperature on the properties as discussed in the introduction. A study of the microstructure of 24S-T after drawing at 450 degrees Fahr. (230 degrees Cent.) revealed no significant changes in grain size. This temperature was sufficiently low and the time at temperature was sufficiently short to prevent major changes in the physical properties.

Effect of Die Temperature on the Drawability of Aluminum Alloys—All of the alloys which were investigated revealed substantial increases in deep drawing properties with increase in die temperature. The influence of temperature on the drawability of the alloys

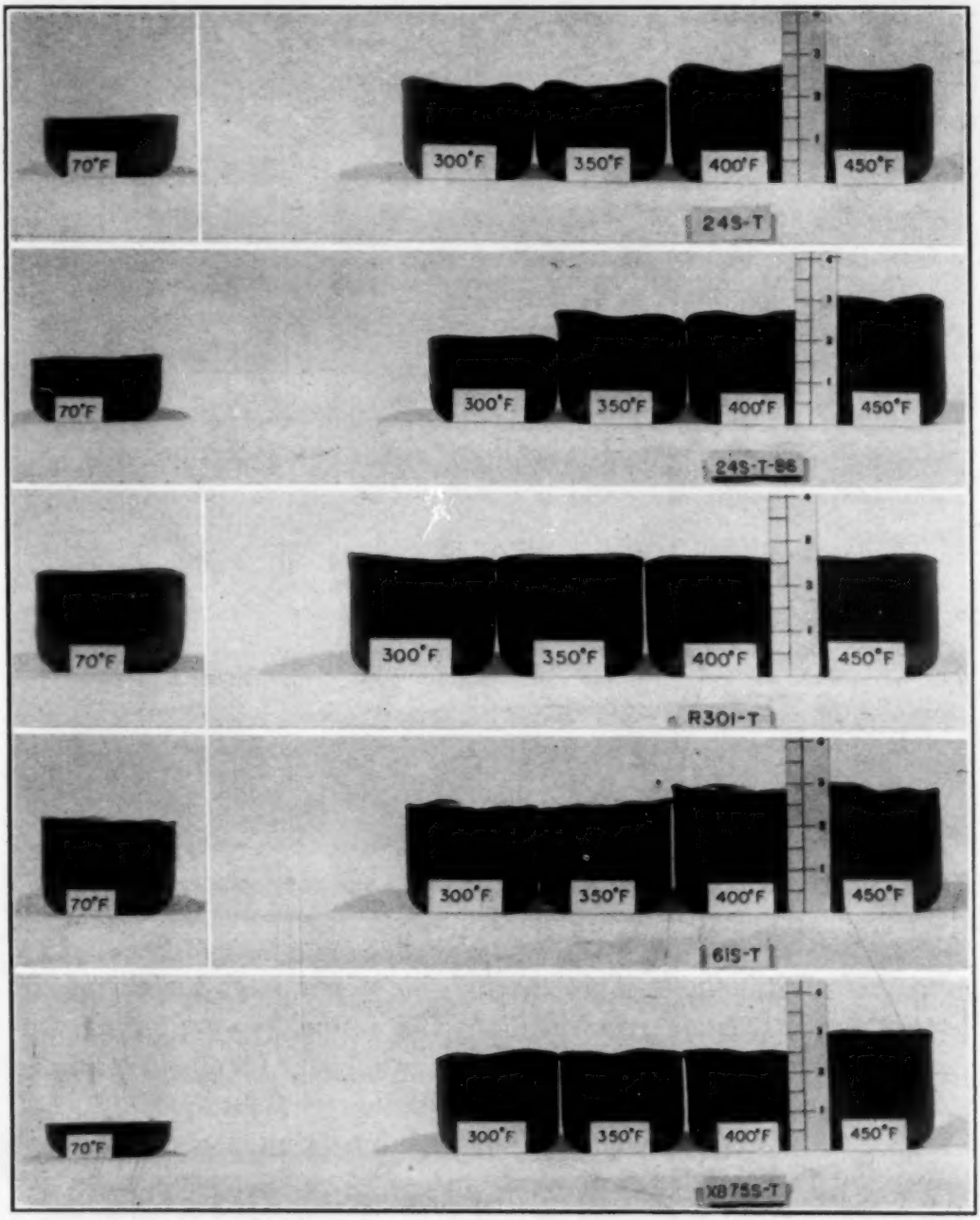

Fig. 10—Maximum Successful Draw at Various Temperatures for Hardened Temper Aluminum Alloys.

investigated is revealed by the photographs in Figs. 9 and 10.

Group 1A alloys, 3S-O and 52S-O, illustrated in Fig. 9, revealed that the highest drawability is achieved at the highest test temperature of 700 degrees Fahr. (370 degrees Cent.). Although these alloys have approximately the same drawability at atmospheric temperatures, 52S-O alloy exhibited superiority over 3S-O alloy at

temperatures above about 500 degrees Fahr. (260 degrees Cent.).

Aluminum alloy 24S-O was the only example of Group 1B alloys examined. As shown in Fig. 9 up to about 400 degrees Fahr. (205 degrees Cent.) there is only nominal improvement in its drawability but at higher temperatures deeper draws may be achieved. The drawability of 24S-O at 700 degrees Fahr. (370 degrees Cent.) is superior to that for 3S-O but slightly below that for 52S-O.

The only example of Group 2A alloys which was investigated was 24S-T. Fig. 10 reveals that the change in drawability of 24S-T is comparatively great for small changes in temperature between 70 and 400 degrees Fahr. (21 and 205 degrees Cent.). It is quite likely that warm drawing of this alloy at temperatures up to about 212 degrees Fahr. may facilitate drawing critical parts which exhibit frequent failures when drawn at 70 degrees Fahr. It is interesting to note that the deep drawing properties of 24S-T alloy at 450 degrees Fahr. (230 degrees Cent.) are as good as the deep drawing properties of Groups 1A and 1B alloys at 70 degrees Fahr. This fact suggests the possibility of deep drawing 24S-T series alloys at 450 degrees Fahr. (230 degrees Cent.) for structural or semistructural applications in lieu of deep drawing 24S-O at 70 degrees Fahr. and then heat treating. By this technique the complications arising from warpage during heat treatment may be overcome and the production steps are simplified.

The general trends obtained for the drawability of the interesting high strength alloys of Group 2B as a function of temperature as shown in Fig. 10 are similar to those for Group 1B. XB75S-T exhibited the poorest drawability of the group at 70 degrees Fahr. but so great was the improvement in drawability with increase in temperature that at 450 degrees Fahr. (230 degrees Cent.) it had a drawability equal to the greatest which was obtained for other alloys in this group. Alloy R301-T exhibited only slight improvement in drawability with increase in temperature. Comparison of the maximum height of cups produced from 24S-T86 and 24S-T shows that 24S-T86 has better deep drawing properties than 24S-T.

In order to provide a basis for design and tooling, the results illustrated in the photographs of Figs. 9 and 10 are plotted in Fig. 11. Both the maximum per cent draw and the drawability are given in these graphs. These charts will also serve as a guide for indicating the possible advantages of warm drawing of parts which frequently fail when drawn at atmospheric temperatures.

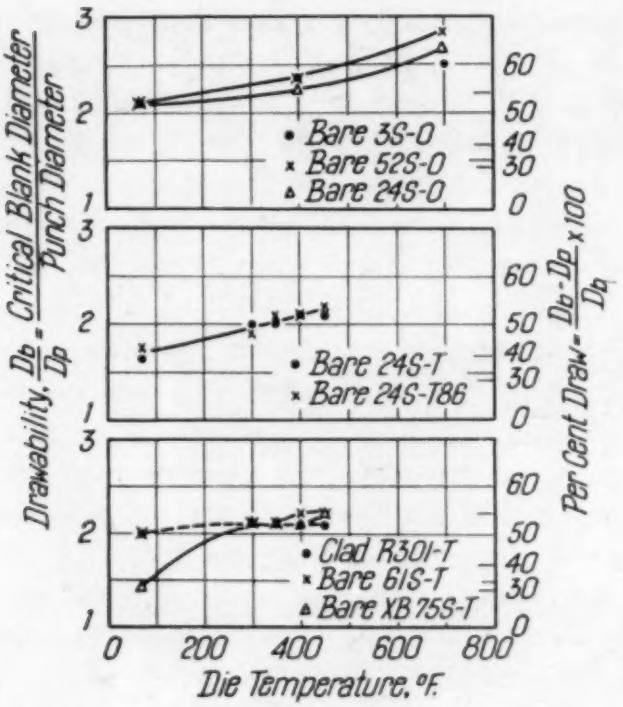

Fig. 11—Effect of Die Temperature on Drawability of Aluminum Alloys. Punch radius = 0.512 inch (8t). Die radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

Relative Cup Height—Conditions of drawing which yield major changes in thickness of the cup walls and conditions of sheet anisotropy which lead to the development of ears on the drawn cup preclude the possibility of interpreting the per cent draw or the drawability in terms of the useful dimensions of the finished product. This difficulty may be surmounted by reporting the deep drawing characteristics in terms of the relative height of the cup.

The effect of temperature on the maximum relative height of the drawn cups is given in Fig. 12. In all of these tests the clearance was sufficiently great to eliminate the effect of ironing on the cup height. Consequently smaller clearances may yield higher cups in all cases where the ironing load does not exceed the tensile strength of the cup walls. The height of the cup was measured from the outside bottom of the cup to the minimum height between the ears. These data, therefore, can be used to establish the maximum permissible trimmed height of a cup.

Effect of clearance on the drawability of aluminum alloys at 450 degrees Fahr. (230 degrees Cent.) is given in Fig. 13. The per cent clearance is defined as:

$$\text{Per Cent Clearance} = \frac{D_d - (D_p + 2t)}{2t} \times 100$$

where D_d = diameter of die
D_p = diameter of the punch
t = original thickness of sheet

Alloys 52S-O, XB75S-T, and R301-T gave greater draws for dies with 4 per cent clearance than for dies with 83 per cent clear-

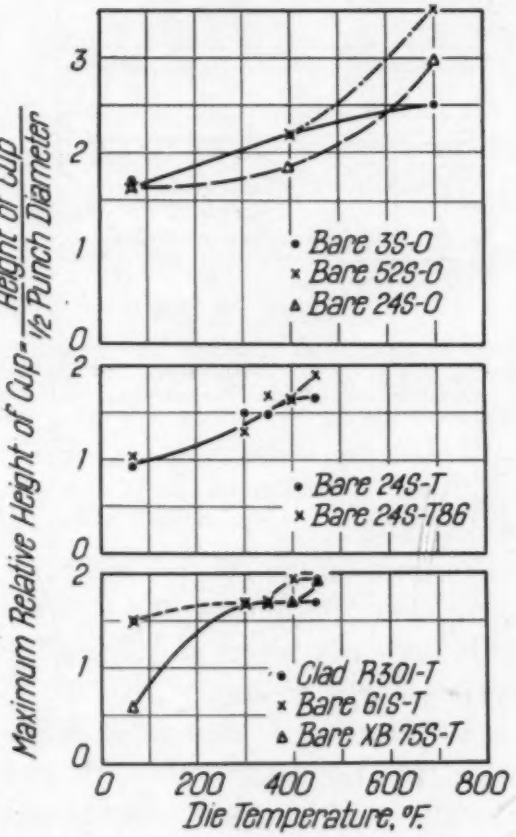

Fig. 12—Effect of Temperature on the Maximum Relative Height of Cup for Aluminum Alloys. Punch radius = 0.512 inch (8t). Die radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

ance. This increase in drawability with 4 per cent clearance dies was due to the ironing out of moderate wrinkles that were not eliminated in an 83 per cent clearance die.

Alloys 24S-T, 24S-T86, and 61S-T showed no change in drawability as a result of changes in die clearance over the range of clearances from 4 to 83 per cent.

For cases in which the cup height is not great enough to cause

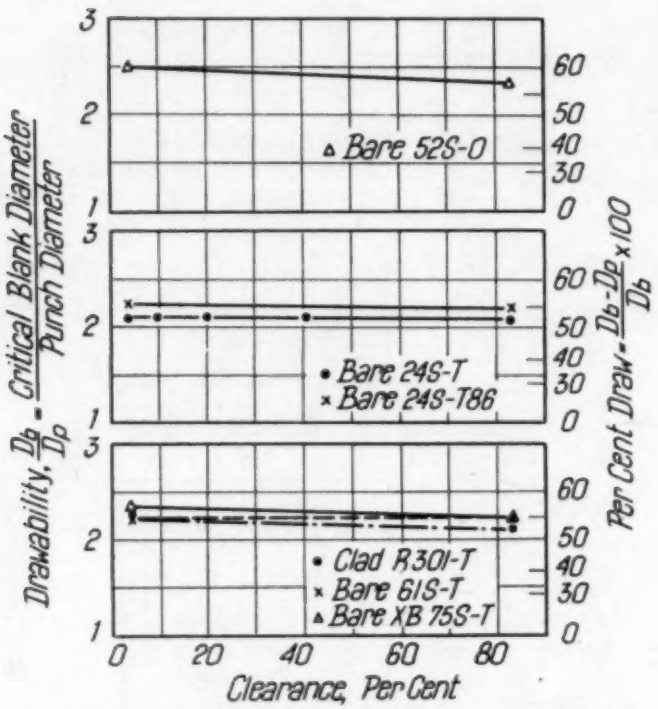

Fig. 13—Effect of Clearance on Drawability of Aluminum Alloys. Die temperature = 450 degrees Fahr. Punch radius = 0.512 inch (8t). Die radius = 0.512 inch (8t).

excessive thickening and consequent heavy ironing, the optimum clearance is near 4 per cent.

The effect of clearance on the relative height of the cup for 24S-T is presented in Fig. 14. Decreasing the clearance over the range of clearances, which does not affect the drawability, results in an increase in the relative height of cup.

Effect of Punch Radius on Drawability—Figs. 15 to 17 present the experimental data for the effect of punch radius on the drawability for all of the alloys which were investigated. The punch radius has a more pronounced influence on the drawability of the hardened alloys than on the drawability of the annealed alloys. The influence of the punch radius, however, is smaller at 450 degrees Fahr. (230 degrees Cent.) than at 70 degrees Fahr.

Effect of Die Radius on Drawability—The drawability of 52S-O and the Groups 2A and 2B alloys as a function of die radius is given in Figs. 18 and 19. In general the drawability increased with increasing die radius. This fact suggests that die radii be made as ample as permissible for deep drawing aluminum alloys.

Speed—Investigations of the effect of speed on the drawability of the precipitation hardened alloys of Groups 2A and 2B were car-

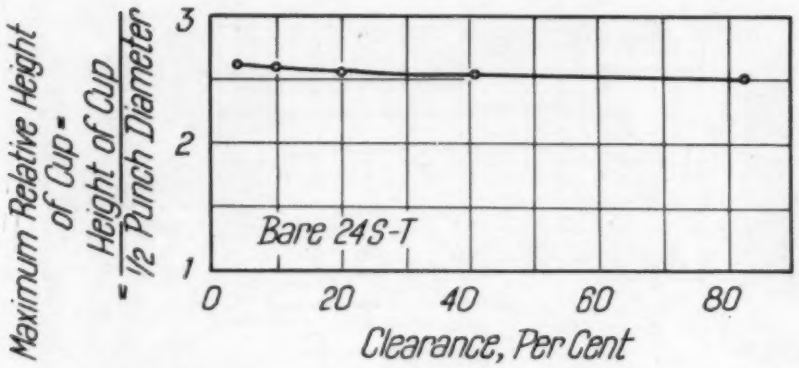

Fig. 14—Effect of Clearance on the Maximum Relative Height of Cup for 24S-T Aluminum Alloy. Die temperature = 450 degrees Fahr. Punch radius = 0.512 inch (8t). Die radius = 0.512 inch (8t).

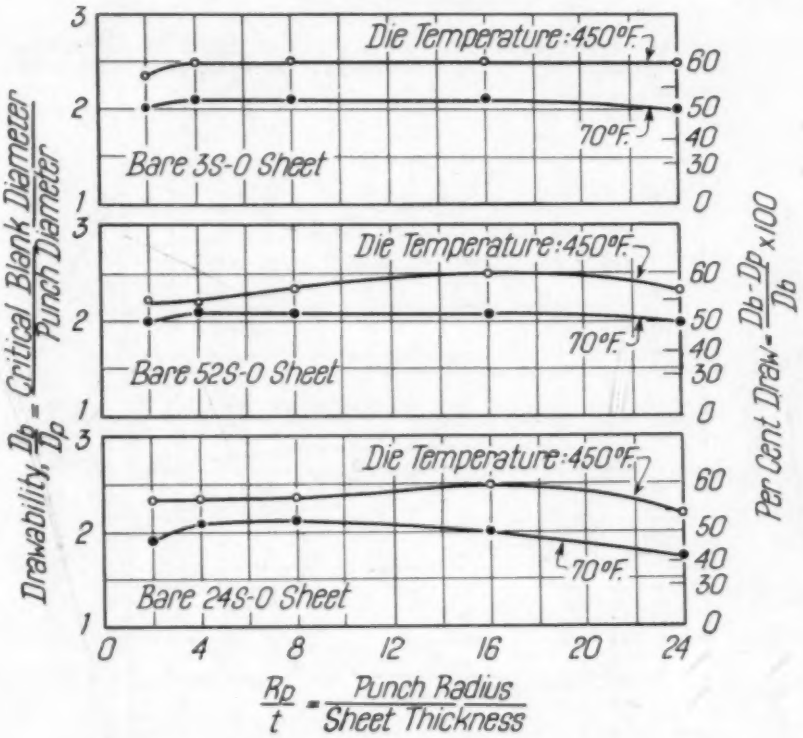

Fig. 15—Effect of Punch Radius on Drawability of Aluminum Alloys. Die radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

ried out at 450 degrees Fahr. (230 degrees Cent.) over a punch speed range of 10 to 145 inches per minute. Alloys 24S-T, 61S-T, and R301-T revealed no change in drawability over the range of speeds used. A small reduction in drawability of 24S-T86 and XB75S-T with increase in speed was observed. The drawability of both of these alloys at 26 inches per minute was 2.22 or 55 per cent but at 145 inches per minute the drawability was 2.10 or 52.5 per cent.

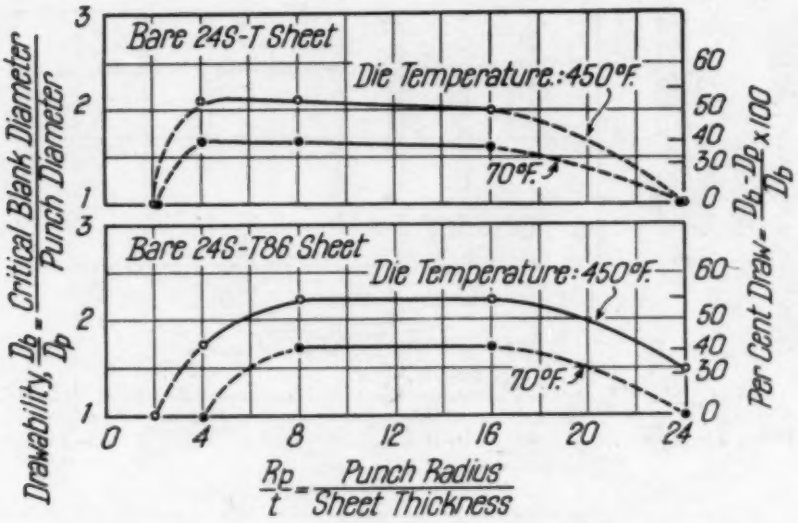
Bare 24S-T Sheet
Die Temperature: 450°F.
70°F.
Bare 24S-T86 Sheet
Die Temperature: 450°F.
70°F.
Drawability, Db/Dp = Critical Blank Diameter / Punch Diameter
Per Cent Draw = (Db-Dp)/Db x 100
Rp/t = Punch Radius / Sheet Thickness

Fig. 16—Effect of Punch Radius on Drawability of Aluminum Alloys. Die radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

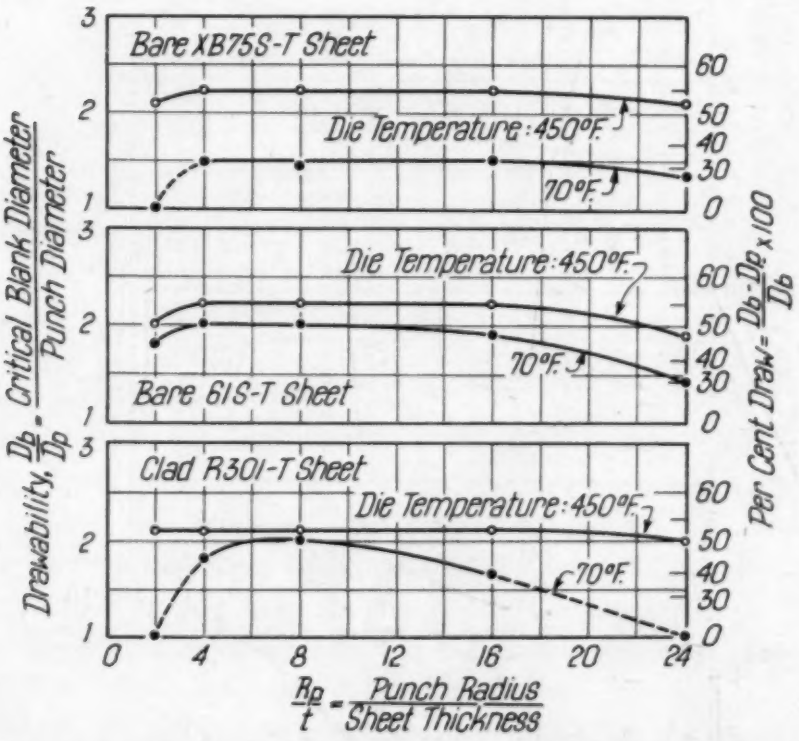

Fig. 17—Effect of Punch Radius on Drawability of Aluminum Alloys. Die radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

Conclusions

1. The deep drawing properties of the aluminum alloys which were investigated improved substantially with increase in temperature.

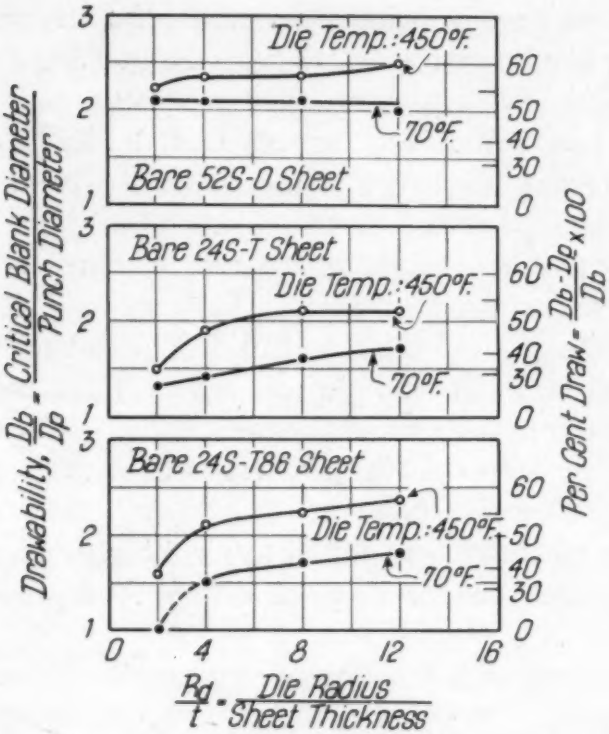
Die Temp.: 450°F.
70°F.
Bare 52S-O Sheet
Bare 24S-T Sheet
Bare 24S-T86 Sheet
Drawability, Db/Dp = Critical Blank Diameter / Punch Diameter
Per Cent Draw = (Db - Dp)/Db x 100
Rd/t = Die Radius / Sheet Thickness

Fig. 18—Effect of Die Radius on Drawability of Aluminum Alloys. Punch radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

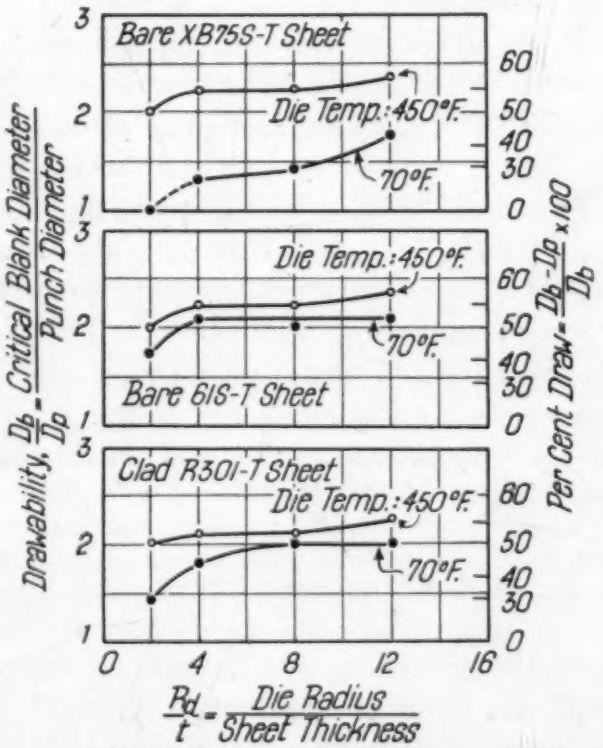
Bare XB75S-T Sheet
Die Temp.: 450°F.
70°F.
Bare 61S-T Sheet
Clad R301-T Sheet
Drawability, Db/Dp = Critical Blank Diameter / Punch Diameter
Per Cent Draw = (Db - Dp)/Db x 100
Rd/t = Die Radius / Sheet Thickness

Fig. 19—Effect of Die Radius on Drawability of Aluminum Alloys. Punch radius = 0.512 inch (8t). Clearance = 83 per cent (no ironing).

2. The maximum deep drawing temperature for the annealed alloys was found to be limited to 700 degrees Fahr. (370 degrees Cent.) due to lubricant failures above this temperature.
3. At 700 degrees Fahr. (370 degrees Cent.) 52S-O exhibited the best deep drawing properties of the annealed alloys; the height of cup which was produced at this temperature was slightly over twice that maximum height which was obtained at atmospheric temperatures.
4. Alloys XB75S-T, 24S-T86 and 61S-T exhibited better deep drawing properties at 450 degrees Fahr. (230 degrees Cent.) than 24S-T and R301-T at 450 degrees Fahr. (230 degrees Cent.) and also better deep drawing properties than the annealed alloys at atmospheric temperature.
5. A 4 per cent clearance was found to yield higher cups than an 83 per cent clearance die for alloys tested at 450 degrees Fahr. (230 degrees Cent.).
6. The optimum punch radius depended upon the alloy and temperature of drawing. In general, higher temperatures permitted the production of cups with sharper bottom radii.
7. In general, the drawability of aluminum alloys increased with an increase in die radius.
8. An increase in the speed of drawing from a stroke rate of 26 inches per minute to 145 inches per minute decreased the drawability of 24S-T86 and XB75S-T slightly at 450 degrees Fahr. (230 degrees Cent.), but had no influence on the drawability of 24S-T, 61S-T, and R301-T at 450 degrees Fahr. (230 degrees Cent.).
9. One of the major factors in the deep drawing of aluminum alloys at elevated temperatures was lubrication. The best lubricant found thus far was a 600W steam cylinder oil modified with one and one-quarter pounds of graphite per gallon of oil. Other lubricants, however, may prove superior and provide better elevated temperature drawing.

Acknowledgments

This report covers part of an investigation sponsored and completely supported by the Office of Production Research and Development of the War Production Board and supervised by the War Metallurgy Committee. The authors wish to acknowledge the guidance

of the WMC throughout this investigation and thank the OPRD for authorizing publication of these data which are based on research done under the "restricted" Project NRC-548, and reported in Report Serial No. W-111.

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7. F. Boulger and F. Dahle, "A Test for Measuring Drawability of Deep Drawing Steels," Battelle Memorial Institute, Columbus, Ohio.
8. Morris Asimow, "An Investigation of the Plastic-Flow Processes Involved in Drawing Cylindrical Shells from Circular Blanks," University of California, Published in *Engineering,* Vol. 3, No. 5, 1936.
9. Greer Ellis, "Detecting Surface Flaws in Non-Ferrous and Non-Magnetic Materials," *The Iron Age,* December 17, 1942.

DISCUSSION

Written Discussion: By L. P. Spalding, chief research engineer, North American Aviation, Inc., Inglewood, Cal.

The authors of this paper are to be congratulated for a very lucid and interesting analysis of a process which, to the writer's knowledge, has not previously been described in technical or trade publications of general circulation. The writer is privileged to have participated in a very minor way in outlining and guiding the very comprehensive program on hot forming aluminum alloys, of which the present paper is a partial report.

In the description of the fundamental mechanism of deep drawing, the existence of sidewall tension stresses and their effect in reducing wall thickness is noted. It would be interesting to know whether or not this reduction in thickness was measured in the tests. This would appear to be of practical importance from the standpoint of strength and dimensional tolerance of finished parts. These measurements might also be significant on the theoretical side since they would have a direct bearing on the magnitude of tension stresses at rupture in those cases where sidewall failures occur. If these data were combined with temperature measurements in the sidewall region and with known elevated temperature tensile strengths, a more complete analysis of such failures should be possible.

The authors' selection of a 5-minute preheating period on the die and their suggestion that this might be reduced in production by furnace preheating deserves more than passing notice. A 5-minute period in the die is of course inconsistent with the production use of draw press operations. On the other hand, if an oven were used it seems quite likely that some pieces in a production lot of blanks might be heated for a considerable time before withdrawal. Since 15 or 30 minutes at the higher temperatures may appreciably affect some alloys this should be very carefully watched. In this same connection the writer believes that the use of a 450 degree Fahr. (230 degree Cent.) drawing temperature on some Group 2 alloys is too high unless the time limit is reduced to about 1 minute. Even 5 minutes' heating is sufficient to cause a 7 per cent drop in the ultimate strength of 75S-T.

The omission of test temperatures between 70 and 300 degrees Fahr. (20 and 150 degrees Cent.) appears to be well justified by the quite definite trends of drawability vs. temperature shown in Fig. 11. Some tests made by the writer's company on stretch forming of 75S-T alloy indicated 250 degrees Fahr. (120 degrees Cent.) as an optimum temperature. However, this observation is not made to contradict the authors' findings, since there is a recognized fundamental difference between deep drawing and stretch forming.

The exclusive use of sawed and deburred annealed blanks, and turned and polished hardened blanks, is to be regretted. In regular production either routing or punch press blanking, with no other finishing operations, is more common. If some spot checks of these methods were made, the data would be quite interesting.

The association of edge fractures with condition of the blanks is common for all types of sheet metal forming. This is quite logical in any stretching operation but seems odd in the present case where circumferential compressive stresses exist at the edge of the blank. This fact suggests that secondary stresses, perhaps imposed by wrinkling, are responsible for the fractures, in which case an increase in hold-down pressure should reduce the number of failures. In this same connection it is important to know whether the edge fractures occurred in the early stages as the periphery of the blank was being drawn in toward the punch, or later as the edges were bent around the die radius. The authors' comments on this point are solicited.

With regard to the alloys tested, it may be noted that the XB75S-T referred to in the paper was an experimental designation for this alloy during its developmental stage. The current commercial product, designated simply 75S-T, has properties identical to the earlier XB75-ST. Incidentally, typical properties of this alloy, which were omitted from Table III, are: tensile yield strength—72,000 pounds per square inch; ultimate strength—82,000 pounds per square inch; elongation in 2 inches—10 per cent. In the same table, the latest data published by the producers[2] show typical properties for 24S-O as follows: tensile yield strength—11,000 pounds per square inch; ultimate strength—27,000 pounds per square inch; elongation in 2 inches—19 per cent.

The reasons for differentiating between Group 1A and Group 1B alloys are not clear. The stated distinction is made on the basis of precipitation hardening at some temperature in the case of 1B alloys (24S-O). However, precipitation

[2]W. L. Fink, J. A. Nock, and M. A. Hobbs, "Aging of 75S Aluminum Alloy", *The Iron Age*, November 1, 1945.

hardening should be negligible or nonexistent in the absence of a solution heat treatment in the case of 24S-O or other Group 1B alloys. The lack of distinction between these groups is further emphasized by the authors' statement that both types may be deep drawn at any temperature up to the point of hot shortness or grain growth. In all tests up to the maximum temperature 24S-O behaved in a manner similar to 3S-O and 52S-O and in general showed properties intermediate between these two Group 1A alloys.

Somewhat in contrast to the stated need for smooth polished die surfaces, the writer's company has found that on some deep drawing jobs better results and less breakage are achieved with punches which are deliberately roughened (for example, by sand blasting) at the nose and radius. The improved results are attributed to elimination of sliding of the metal over the punch radius which causes excessive stretching in the cup bottom instead of drawing out the sidewalls. These observations are largely confined to room temperature but it is believed that they would be equally applicable to elevated temperature work. This technique of roughening the punch should be more useful on nonsymmetrical sections than on circular cylindrical cups due to the greater importance of controlling and distributing metal flow.

Although the grain size of the sheets used was determined for Group 2 alloys and it is noted that there was no significant change in grain size for the heating cycles used, there is no correlation shown between this property and the drawability attained or the occurrence of failures. The bend cracks illustrated in Fig. 7a are thought to represent an extreme case of the phenomenon of "orange-peel" which has been associated with abnormally large grain size.

The writer would like to know if any tests were made with clad sheets having the same base composition as some of the alloys reported here. Based on production experience with deep drawing at room temperature, and also with other forming methods, substantially better performance would be expected of clad alloys because of the high ductility and flow characteristics of the coating.

The authors indicate that other phases of their work on elevated temperature forming and properties of aluminum alloys may be published. Being quite familiar with the available material, the writer sincerely hopes that this will be done in the near future so as to make this wealth of information more generally available than was possible under war-time restrictions.

Written Discussion: By C. F. Nagel, Jr., chief metallurgist, Aluminum Company of America, Pittsburgh.

This paper contains data which should be of practical value. It is apparent that the authors planned the tests with due consideration to avoid disturbing influences and carried out the tests very carefully. Consequently, the resulting data can be relied upon.

The Aluminum Company of America has conducted similar tests on the same alloys reported upon in the referenced paper. The results obtained in these tests agree very well with the data presented in the referenced paper.

On page 265, when commenting upon Group 1A, it is stated that alloys in this group may be deep drawn at any temperature up to their hot short or grain growth temperature without introducing significant changes in properties. This would mean that when starting with annealed alloys, at room temperature, the sidewalls of the drawn shell would still be in the annealed condition. This is

known not to be so. With a reasonably deep draw, it is known that the metal in the sidewalls of the shell will be roughly equivalent to the quarter-hard temper and, if a number of successive draws are taken, the metal in the sidewalls may reach the full-hard state. This must necessarily be true because, in the act of drawing a shell, the metal is cold-worked and this must harden the metal and increase the strength.

On page 270, in the paragraph "Earing", it appears that the authors believe the increased amount of earing observed, when drawing at increased temperatures, is due to the increase in temperature of the metal. It is believed by us that this greater amount of earing, observed at elevated temperatures, is not due to the fact that the metal temperature was higher, but rather that, with increase in metal temperature, a deeper shell can be drawn. It is suspected that if shells of the same depth were drawn, but with metal at different temperatures, the amount of earing would be found to be substantially the same. It has been our experience that the amount of earing is a matter of metallurgical structure of the metal.

On page 268, mention is made of the fact that, at the higher temperatures, the oil residue charred and that this required cleaning the parts after each draw. While this would be possible in practice, it is believed that in production it would be found simpler and more certain to put in an extra draw in order to get the desired height of the shell without introducing such a cleaning operation.

The resulting surface, after the draw, may be very important and is definitely so when finishing operations must be performed in order to obtain a highly polished finish. Galling of the metal is more likely to occur at elevated temperatures. Thus, one must always weigh the advantages of obtaining a deeper draw, by means of raising the temperature of the metal, against the possibility of increasing the finishing operations. When the finished article must be highly polished, it will usually be found that if marring of the surface is encountered at elevated temperatures, thus increasing the amount of surface grinding required, it will be less expensive to drop the metal temperature and introduce an additional draw.

Written Discussion: By E. H. Burkart, group engineer, Metals Research, Lockheed Aircraft Corp., Burbank, Cal.

The authors have presented considerable information on deep drawing of aluminum alloys at elevated temperatures, which will be very useful in part and tool design to facilitate production.

As stated, the selection of maximum temperature for drawing of the various materials was based exclusively on the physical effects of temperature, and the investigation of other factors, such as corrosion, endurance, etc., is indicated. In regard to corrosion, the hot forming of 24S-T in the range of temperature 300 to 450 degrees Fahr. has resulted in a markedly increased susceptibility of this material to intergranular and stress corrosion. Data obtained in recent tests also show an appreciable loss in the strength properties of 24S-T after heating at these temperatures for a short period of time (½ to 2 minutes). Consequently, it is felt that hot deep drawing of 24S-T should be undertaken only when the part is to be subsequently aged at elevated temperature for the prescribed period (375 degrees Fahr. for 10 ±½ hours).

The use of a lubricant modified with graphite is undesirable from the production standpoint due to the difficulty encountered in the recovery of the degreasing

agent, and further investigation should be made to obtain a more suitable liquid lubricant.

Written Discussion: By Francis W. Boulger, assistant supervisor in metallurgy, Battelle Memorial Institute, Columbus, Ohio.

The authors are to be congratulated upon their careful and systematic investigation of the effect of temperature on the drawability of aluminum alloys. Did the cups drawn at elevated temperatures have as good a surface finish as those drawn at room temperature? The improvement in drawability of aluminum alloys at slightly elevated temperatures agrees with results obtained by others on zinc alloys and suggests that tests on other materials might be worthwhile.

The data emphasize the inadequacies of tensile tests and show the possibilities of cup drawing tests in evaluating the drawing quality of sheet materials. From the latter standpoint, it would be well to elaborate on a few items not discussed at length in the paper. Did the authors find that different lubricants or different clearances affected the limiting per cent draw in room-temperature tests? The paper gives the impression that the lubricants were unsatisfactory only when they broke down at the testing temperature. The effect of clearance on test performance was only described for tests at 450 degrees Fahr. (230 degrees Cent.). If the per cent draw at room temperature was not appreciably affected by either variable, it would suggest that reliable ratings could be based on tests under one set of standard conditions.

The anomalous effects of speed and clearance on drawability of the alloys at 450 degrees Fahr. (230 degrees Cent.) deserve further consideration. Alloys 24S-T, 61S-T, and R301-T gave the same results in cup drawing tests for punch speeds ranging from 10 to 145 inches per minute. The reduction in drawability of 24S-T86 and XB75S-T with the highest speed in this range was very small and equalled one division on the scale employed. Similarly, an increase in clearance from 4 to 83 per cent apparently lowered the per cent draw values one division on the rating scale for four of seven materials tested at 450 degrees Fahr. (230 degrees Cent.).

The number of disks tested for each condition is not stated. Did the blanks give 100 per cent success or 100 per cent failure for each testing condition? If not, then the indicated effects of speed and clearance may have resulted from the discontinuous scale employed. In cup drawing tests made by the writer, on steel, samples from the same sheet often varied in performance so that it was necessary to use breakage figures in rating drawing quality. For example, the first blank for two different steels might indicate that both materials had the same drawability under fixed test conditions. If additional blanks were tested, the data sometimes showed 90 per cent breakage for one steel and only 20 per cent for the other. Obviously the second steel was slightly superior, but the ratings assigned on the basis of fewer tests would vary, depending upon which blanks happened to be tested first. Such behavior might explain the drop in drawability of one division, on the authors' scale, noted for increasing speeds on two alloys and for the larger clearance on some of the materials tested.

Written Discussion: By L. Schapiro, Douglas Research Laboratories, Douglas Aircraft Co., Inc., Santa Monica, Cal.

A paper presenting the large volume of experimental work and the detailed care in performance and presentation of results such as these authors

have given us here merits the commendation "Congratulations—well done." Industrial users of aluminum alloys in deep drawing operations have been furnished in this paper ample experimental evidence by which to be guided in their work.

In their introductory remarks the authors have outlined the combinations of stresses imposed in the cup drawing operation and have also pointed out the factors which thus far keep the drawability problem in the experimental stage. We can but look forward to the time when the ductility of the various aluminum alloys has been evaluated for various stress ratios and at various temperatures, and our various forming operations have likewise been evaluated for the stress ratios applied in each, so that a combination of these two areas of knowledge can permit the prediction of performance in an old or a new sequence of operations. A classification of materials based on ductility in several stress systems permits an adequate selection for a forming operation involving one or more of the stress systems. The authors' comments about the edge condition of the test blanks may be considered a slight understatement. For the Group 2 alloys, our experience with forming operations indicates edge condition to be a very important factor in utilizing the full ductility of these alloys.

Written Discussion: By K. R. Jackman, chief test engineer, Consolidated Vultee Aircraft Corp., San Diego, Cal.

The authors are to be congratulated on their excellent paper covering a field in which there has been very little information. As the authors mentioned, they expect to determine further the effects of the elevated temperatures on the mechanical properties of the drawn cups discussed in their paper. However, it might be pertinent to the discussion of the paper at the meeting to present some limited data obtained at Consolidated Vultee on the effects of elevated temperatures on the mechanical properties of several of the aluminum alloys.

It has been found that 24S material in the heat treated and aged condition does not lose appreciably in mechanical properties after exposures of 5 to 10 minutes at 400 to 425 degrees Fahr. (205 to 220 degrees Cent.), nor do the zinc-bearing alloys, 75S and R303, show an appreciable reduction in mechanical properties when heated at 350 to 375 degrees Fahr. (175 to 190 degrees Cent.) for short periods. From these data it might be deduced that part of the good drawability of the aged materials could have been attributed to the annealing effect of the elevated forming temperatures, particularly the 450 degree Fahr. (230 degree Cent.) and higher forming tests.

Authors' Reply

The authors wish to express their appreciation for the expert advice furnished to them by the Western Aircraft War Production Council and the War Metallurgy Committee in conducting this research. Although it would be impossible to list all persons who contributed to this study, it is appropriate to give special acknowledgment to M. Nelles, onetime chairman of the Testing and Research Panel of the A.W.P.C. and later Deputy Chief of the Minerals and Materials Branch of the O.P.R.D. who instituted the elevated temperature forming investigations, and to L. L. Wyman, Research Supervisor of the W.M.C., under whose direction this work was co-ordinated. It is indeed gratifying to note, among the discussers, Messrs. Spalding, Schapiro, Burkart, Boulger, Jackman, and Nagel

who, as members of the W.M.C. or the A.W.P.C. and representative companies, participated in the early discussions and assisted in formulating plans for this investigation. Their comments and expert opinions serve to extend the value of this paper. No attempt will be made to reply to the individual discussers inasmuch as many of the comments and questions overlap and therefore may adequately be treated at one time.

No intensive study of the reduction in wall thickness was attempted in this survey which was primarily directed toward revealing the gross differences in the drawability of a wide range of alloys at several temperatures. The sidewall thinning immediately above the bend radius of the drawn part is known to be influenced by a number of factors among which are the per cent draw, the extent of ironing, and (for the hot die-cold punch arrangement) the rate of drawing. Greater per cent draws and greater amounts of ironing usually result in greater amounts of sidewall thinning; slower draws result in greater cooling of the sidewalls and consequently less thinning. Quantitative studies on sidewall thinning, however, were not made.

Although edge fractures were not common for the range of alloys and temperatures employed, their existence has interesting implications. A few spot tests appeared to indicate that they were produced when the blank edge was drawn around the bend radius. Their presence, however, does not necessarily imply that tensile stresses are present for many metals are now known to fail in accordance with a Maximum Shear Stress Law although a Maximum Tensile Stress Law may be valid for other metals. If the Maximum Shear Stress Law is the determining criterion for failure it is possible to obtain fracture under pure compression for which the maximum shear stress lies on planes at 45 degrees to the compression axis and has a value equal to one-half of the compression stress.

Roughening the die can only have deleterious effects in increasing friction (and thus the load that the sidewalls must sustain) and in increasing scoring when deep drawing cups. Spot roughening dies is usually practiced to restrain metal flow when drawing asymmetrical objects. Reduction of corner cracks in boxes may be achieved by employing roughened die radii maintaining a high polish at the corners, thereby distributing the load more uniformly over the wall. Roughened punches were not used because an attempt was made to secure comparable data for all types of punches employed; with the facilities available a uniform polish on all punches was more easily maintained than uniform roughness.

No tests were made on alclad material in order to limit the scope of the investigation. It was believed that the results obtained for bare stock in general would be conservative when applied to clad material.

The designation XB75S-T was retained in the present report because the stock employed was produced during the final stages of development of 75S-T alloy. Since XB75S-T was experimental, typical values could not be quoted. Although the equivalence of XB75S-T and 75S-T is now established it was considered advisable to retain the experimental designation in order to accurately classify the stock used.

The major reason for dividing alloys in the annealed condition into subgroups 1A and 1B arises from the fact that some, classified 1A, are not precipitation hardenable whereas the remainder, 1B, are precipitation hardenable. Usu-

ally 1A alloys are used as formed, whereas the 1B alloys are commonly solution treated and precipitation-hardened after forming in order to develop optimum properties.

The classification of the alloys into various groups was based on their response to high temperatures rather than their atmospheric temperature work hardening characteristics. Although the sidewalls of cups drawn at atmospheric temperatures will be work-hardened, the bottoms, which are not severely worked, will substantially retain their initial properties. Consequently the classification is valid also for atmospheric temperature.

The cost of the numerous blanking dies which would have been required for the present study and the limited number of blanks required suggested the use of sawed blanks for this preliminary survey. It is possible that the more severe edge working resulting from blanking may induce more frequent occurrence of edge cracking than was observed in this investigation. In preliminary tests on some of the Group 2 alloys it was noted that edge cracks occasionally developed. When the edges were machined and polished greater draws could be achieved before edge cracking again limited the draw. Machined and polished edges were then adopted to provide a more uniform basis of comparison.

In general the surface condition of a hot-drawn part was slightly inferior to a cold-drawn part. Particularly severe scoring was obtained under conditions leading to rupture of the film of lubricant and die pickup of the drawn metal. Development of readily removable, stable, nontoxic liquid lubricants or soluble solid lubricants for hot forming will play an important part in the future extension of hot forming techniques. No intensive survey of lubricants was made in the present study.

Only evaluation of gross significant differences in drawability was attempted in this survey. Consequently no statistical study was made to survey blanks covering the range of mill stock that varies over the specification range or variations that may exist in blanks from the same sheet. At least eight blanks were drawn in the critical range. Generally the four larger blanks failed and the four smaller ones drew successfully because of the large interval between adjacent blank sizes. Occasionally, however, the critical size was very close to one of the blank sizes resulting in several failures and several successes for that size. In this event (for blanks from the same sheet) all blanks one step higher in the scale fractured whereas all blanks one step lower in the scale drew successfully. Whenever the results appeared to be anomalous, additional tests were performed. Spot tests on hot deep drawing 1-inch diameter cups on a variable speed crank press using a water-cooled punch have shown that the drawability decreases with increasing speed. This effect, however, may result from the higher temperatures retained by the cup wall as it is drawn at higher speeds, rather than a pure speed effect which would be obtained when the die, punch and blank are initially at the same temperature.

The interest in more theoretical analyses of the deep drawing process is encouraging. Although the theories of plastic flow of work hardenable metals, heterogeneous flow, and fracture are yet incomplete, much progress has been made. A fairly accurate analysis of the drawn flange is now possible but the drawing around the die radius has not yet been satisfactorily solved. When this is done it should be possible to establish the relative merit for deep drawing

cylindrical cups. Although the analytical approach for drawing this simple object appears to be immediately imminent, the authors believe that intensive empirical studies will for some time continue to be necessary for a complete description of deep drawing boxes and other more complicated shapes.

Earing of circular cups is a reflection of the anisotropy of the sheet being drawn. The extent of earing commonly increases with increase in the per cent draw. If, however, the drawing temperature is altered the mechanism of single crystal deformation may also change in such a manner as to modify the extent, number, and orientation of the ears. The authors did not attempt to determine whether the changes in earing recorded in Table IV as a function of temperature were due to the larger draws that were achieved at the higher temperatures or whether the temperature per se was an influencing factor. It is interesting to note, however, that increased hold-down forces, especially at the higher temperatures, caused increased earing for draws that were otherwise identical.

DEEP DRAWING ALUMINUM ALLOYS AT ELEVATED TEMPERATURES

Part II—Deep Drawing Boxes

By Dan M. Finch, Scott P. Wilson and John E. Dorn

Abstract

The present investigation was conducted to determine whether the improvement in deep drawing at elevated temperature reported previously for cylindrical cups is obtainable for box-type parts.

The results obtained for a limited number of alloys and two temperatures, 70 and 450 degrees Fahr., illustrate the decided advantage of deep drawing box-type parts of aluminum alloys at elevated temperature. For all of the alloys tested the maximum height of the box drawn at 450 degrees Fahr. was greater than the maximum height of box drawn at 70 degrees Fahr. For the T temper alloys the square boxes drawn at 450 degrees Fahr. were more than twice the height of those obtained at 70 degrees Fahr.

The improvement in drawability at 450 degrees Fahr. as compared to 70 degrees Fahr. was the greatest for the high strength aluminum alloys. The results warrant consideration of deep drawing aluminum alloys in the hardened temper at elevated temperatures rather than deep drawing the alloys in the annealed condition at room temperature and subsequently heat treating and aging.

The improvement in drawability at elevated temperatures also indicated that the application of this technique to the drawing of parts from standard alloys will result in a reduction in the number of press and anneal operations required by reducing the number of re-draws.

PREVIOUSLY reported investigations[1] on the deep drawing properties of aluminum alloys revealed that elevated temperature drawing frequently permits substantially greater draws than may be

[1] "Deep Drawing Aluminum Alloys at Elevated Temperatures, Part I—Deep Drawing Cylindrical Cups," published in this volume of Transactions, pages 254 to 289.

A paper presented before the Twenty-seventh Annual Convention of the Society held in Cleveland, February 4 to 8, 1946. Of the authors, Dan M. Finch is research engineer, Scott P. Wilson is engineer, and John E. Dorn is associate professor, University of California, Berkeley, California. Manuscript received March 12, 1945.

achieved by drawing at atmospheric temperatures. The benefits of elevated temperature drawing were evaluated for the simple operation of deep drawing cups. The present investigation was planned to determine if these benefits also apply to the deep drawing of box-type parts. This paper is based on work done under Project NRC-548 for the Office of Production Research and Development of the War Production Board under the supervision of the War Metallurgy Committee and with the co-operation of the aircraft industry.

Table I
Aluminum Alloys Investigated

Group	Alloy	Surface
1A	.....	....
1B	24S-O	bare
	75S-O	bare
2A	24S-T	bare
2B	24S-T81	bare
	24S-T86	bare
	75S-T	bare

The aluminum alloys selected for study are given in Table I according to the classification previously adopted in Part I of this report series. In order to reduce the number of tests, the forming temperatures studied were limited to atmospheric temperature and 450 degrees Fahr. Although higher forming temperatures can be used for deep drawing the Group I alloys, 450 degrees Fahr. (230 degrees Cent.) was selected inasmuch as the general aging characteristics of the Group II alloys preclude the use of higher temperatures which cause a permanent decrease in yield strength accompanying overaging.

Experimental Equipment and Technique

The Draw Press—The equipment used was a subpress assembly consisting of an electrically heated die plate, variable pressure heated hold-down pad, and a water-cooled punch assembled in a steel frame and mounted in a variable speed hydropress. The temperature of the die plate and hold-down pad was maintained by controlling pyrometers to within ± 8 degrees of the 450 degree Fahr. (230 degree Cent.) setting. The temperature variation across the die plate was determined by using a dummy specimen in which thermocouples were imbedded. The maximum variation was within ± 4 degrees Fahr. of the mean temperature.

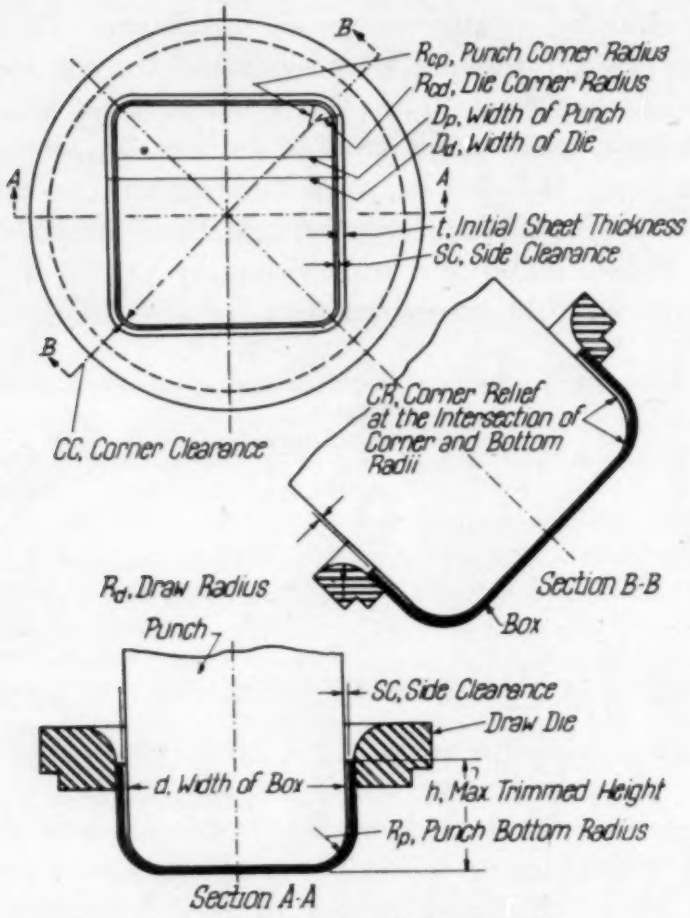

Fig. 1—Illustration of Terms Used.

Punch and Die Design—The important design variables for square box punches and dies as illustrated in Fig. 1 are:

D_p = Width of punch across flats
D_d = Width of die across flats
d = Width of box (inside width, taken as width of punch)
R_d = Draw radius
R_p = Punch bottom radius
R_{cd} = Corner radius of die
R_{cp} = Corner radius of punch
CR = Corner relief
CC = Corner clearance
SC = Side clearance

The terms "corner relief" and "corner clearance" are not clearly defined in the literature (1), (2), (3)[2] but as used in this report are defined as follows:

Corner Clearance is the distance between the corner wall of the punch and land of the die minus the sheet thickness measured along a line bisecting the corner angle.

[2]The figures appearing in parentheses pertain to the references appended to this paper.

Corner Relief is the term applied to the modification of the transition surface formed at the intersection of the corner and bottom radii of the punch by fairing in these curves. Usually metal is removed in the region of the corner intersection just sufficient to distribute the drawing load and increase the corner bend radius so that larger draws can be achieved without having failures develop in this critical area. The extent of the fairing depends upon the radius used at the corner of the box. The greatest corner relief is required for small corner radii.

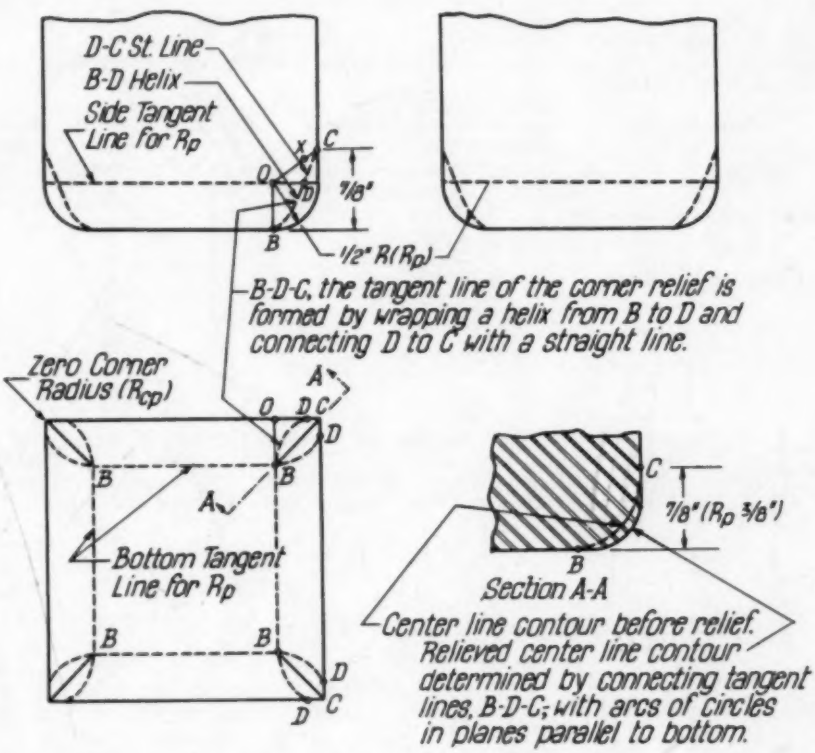

Fig. 2—Illustration of the Method Used to Provide Punch Corner Relief.

The corner relief selected for the punches in this report proved to be satisfactory. The amount of corner relief varied since the punches had variable corner radii. The corner relief was a maximum for the sharp (zero) corner radius punch and was zero for the punches with corner radii 0.50 inch or larger, since for such punches the bottom radius was continuous around the punch nose.

The method used to generate the corner relief surface for the punches used herein is illustrated in Fig. 2. A template was made for the curve along the centerline, BC, of the relief section. The centerline curve is established by projection or lay-out methods. The

two BDC lines in each corner establish the points of tangency for circular arcs drawn in planes parallel to the bottom of the punch. The intersection of these circular arcs with the centerline BC establish the curve for the template used in hand forming the relief surface.

Line BDC is determined by a flat sheet wrapped over the end of

Table II
Die Rings and Punches

(For 0.064 sheet)

Die Rings
D_d = 3.120 in. across flats
SC = 3% clearance
R_d = 0.500-in. draw radius

Die Ring No.	Corner Radius R_{ed}	R_{ed}/t	Corner Radius / Width of Box = R_{ed}/d
B1	1/16"	1	0.021
B2	1/8	2	0.042
B3	3/16	6	0.125
B4	5/16	10	0.208
B5	1 1/8	18	0.375
8	3.133	Round	0.500

Punches
D_p = 2.984 in. across flats
R_p = 0.500-in. bottom radius

Punch No.	Corner Radius R_{ep}	R_{ep}/t
B1	0"	0
B2	1/16	1/2
B3	1/32	1
B4	3/32	1 1/2
B5	1/8	2
B6	5/32	5
B7	11/64	5 1/2
B8	3/16	6
B9	13/64	6 1/2
B10	5/16	10
B11	1 1/8	18
5	1 1/2	24 (round)

Punch and Die Combinations for Various Corner Reliefs

Die No.	Die Corner Radius R_{ed}	Punch Number	Punch Corner Radius R_{ep}	Side Clearance Per Cent	Actual Corner Clearance Per Cent
B1	1/16"	B1	0"	5	18
B1	1/16	B2	1/32	5	27
B1	1/16	B3	1/32	5	47
B1	1/16	B4	3/32	5	67
B2	1/8	B5	1/8	6	50
B3	3/16	B6	5/32	7	8
B3	3/16	B7	11/64	7	28
B3	3/16	B8	3/16	7	49
B3	3/16	B9	13/64	7	69
B4	5/16	B10	5/16	11	52
B5	1 1/8	B11	1 1/8	12	49
8	1 1/2	5	1 1/2	7	7 (round)

the punch with one point on the straight edge of the sheet at B and the other point at C. As the corner radius increases from zero the tangent line BDC moves so that the end point x travels along CD. The end x is at the intersection of C-O and the tangent to the corner radius.

Interchangeable die inserts and punches for a series of 6 corner die radii ranging from 1/16 inch to 1 1/2 inches (round die) and for variable corner clearance using 1/16-inch and 3/8-inch corner die radii were machined from tool steel, hardened and polished. The dimensions of

these dies and punches are presented in Table II along with the corner clearance obtainable with various punch and die combinations. The nominal design side clearance of 5 per cent at 70 degrees Fahr. between the sides of the punch and die varied betwen 5 and 12 per cent among the actual dies. It was observed that clearance variation within these limits did not influence the drawability.

Blank Preparation—A series of blank sizes using the per cent draw as a basis was calculated for every 2½ per cent increase from 20 to 70 per cent draw. This series of blanks is presented in Table III.

Table III
Draw Blank Sizes

(Each specimen was stamped with the direction of rolling, alloy, sheet number and diameter. All blanks were 0.064 inch thick.)

Per Cent Draw	Blank Diameter (D_b) Inches
20	3.75
22½	3.88
25	4.00
27½	4.14
30	4.29
32½	4.44
35	4.62
37½	4.80
40	5.00
42½	5.22
45	5.45
47½	5.72
50	6.00
52½	6.31
55	6.66
57½	7.06
60	7.50
62½	7.00
65	8.57
67½	9.23
70	10.00

For the O temper alloys the blanks were prepared by sawing to size on a metal band saw and deburring the edge. The blanks for all the T temper alloys were first sawed oversize and then machined on a lathe to size, leaving a smooth edge. This latter procedure was used on the T temper alloys to eliminate the effect of edge conditions which had an important influence on the lip fracture type of failure frequently encountered. A few special runs were made using developed blanks.

Standard Test Conditions—Throughout the series of draws for determining the effect of corner radius and temperature on drawability other conditions were maintained constant. Round blanks were cut from 0.064-inch bare, sheet aluminum. The specimens

were oriented with the direction of rolling at 45 degrees to the sides of the box. All boxes had 3-inch-square inside dimensions. The punch bottom radius and draw radius were both 0.50 inch or approximately 8 times the sheet thickness. The side clearance varied due to machining tolerances between 5 and 12 per cent of nominal sheet thickness. Equal punch and die corner radii giving a corner clearance of 48 per cent were used for the investigations of effect of corner radius on drawability.

Water cooling the inside of the punch kept the outside wall temperature between 70 and 80 degrees Fahr. depending on the surrounding temperature. A punch speed of 26 inches per minute was used. The specimens were lubricated with a mixture of 1 pound of powdered graphite and 1 gallon of 600W steam cylinder oil. Preheating the specimens was done on the hot die plate by holding unheated specimens under the hold-down pad 5 minutes before drawing.

Results and Discussion

Measures of Drawability

The significant measures of the drawability of metals are those that reveal the maximum successful draw that may be achieved in a single operation. The two measures of drawability considered in this report are the h/d ratio and the per cent draw. The h/d ratio is defined by:

$$\frac{h}{d} = \frac{\text{Height of Box}}{\text{Width of Box}}$$

where h = Height of Box = actual outside dimension from bottom of box to lowest point on top edge of box (maximum trimmed height)

d = Width of Box = inside width

The per cent draw is defined by:

$$\text{Per Cent Draw} = \frac{(D_b - d)}{D_b} \times 100$$

where D_b = Blank Diameter
d = Width of Box = Inside Width

The maximum h/d ratio presents a measure of the maximum final height of box for a given set of drawing conditions, whereas the per

cent draw gives a measure of the maximum blank size that can be drawn for the same conditions. Generally the height of cup is a more informative measure for design and production use, although both are required if an actual part is to be formed since a size of blank must be selected that can be drawn to the desired height.

Types of Failure

A number of types of failures were encountered in deep drawing square boxes from aluminum alloy sheet as is shown in Figs. 3 and 4. The peaks on square boxes are due to the fact that square boxes were produced from circular blanks. This is in contrast to the ears formed

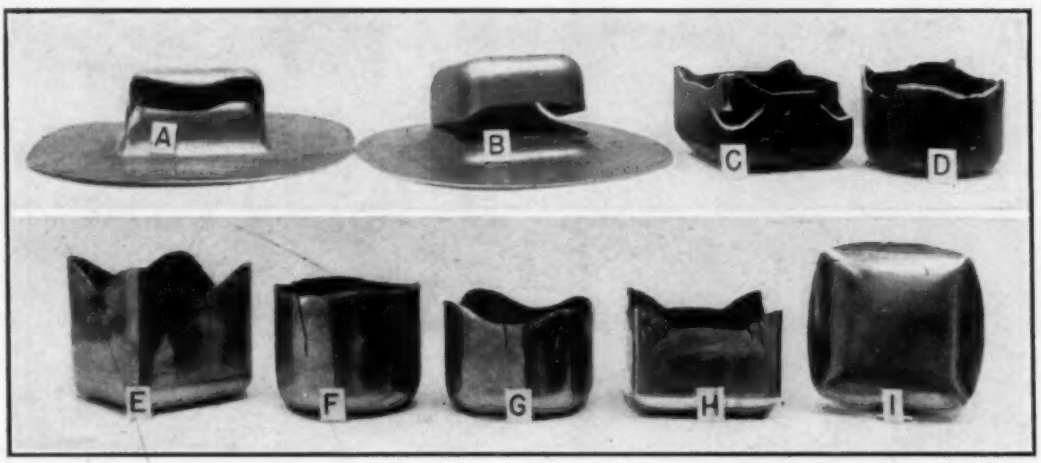

Fig. 3—Types of Failures. A—Bottom radius tear. B—Wall tear. C—Corner wall tear. D and E—Ironing tears. F, G and H—Lip fractures. I—Corner bend cracks.

on circular cups drawn from round blanks which are due to sheet anisotropy.

A. *Bottom Radius Tear*—The bottom radius tear was a frequent type of failure encountered in drawing cylindrical cups at 70 degrees Fahr. Failure occurred when a high drawing force caused local necking around the radius with consequent tearing. The origin of a bottom radius tear occurring in boxes with small corner radii is believed to be at the intersection of the bottom radius, R_p, and the corner radius R_{cp}, where severe thinning was observed on several of the drawn boxes. Corner relief is an important factor in reducing the failures in this region.

B. *Wall Tear*—The wall tear shown in Fig. 3B occurred when the corner and bottom punch radii were sufficient to prevent failure in the bottom of the cup. Failure took place when the drawing force

was greater than that which could be carried by the walls. These tears occurred more frequently at 450 degrees Fahr. (230 degrees Cent.) using a water-cooled punch than at 70 degrees Fahr., since the lower portions of the walls around the punch radius are cooled and thereby strengthened by the cold punch.

C. *Corner Wall Tear*—Corner wall tears appeared most frequently when drawing O temper alloys at 70 degrees Fahr. and occasionally when drawing the T temper alloys at 450 degrees Fahr. (230

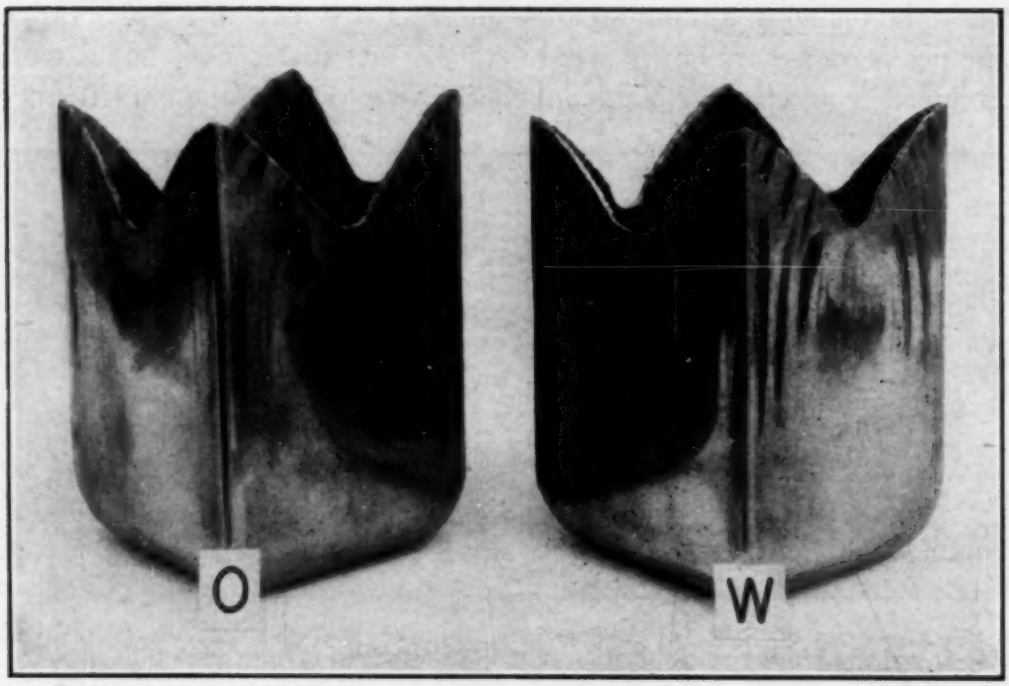

Fig. 4—Wrinkle Examples. O—Slight wrinkle—acceptable. W—Moderate wrinkle —not acceptable.

degrees Cent.). This type of failure occurred in one or more corners of the box as a result of severe flow conditions and high drawing forces concentrated in the corners. Not all of the failures were brittle as shown in Fig. 3C but occasionally revealed considerable local deformation.

D and E. *Ironing Tears*—Thickening of the wall of the box during the draw may exceed the clearance. Ironing starts when the distance between the punch and the die is completely taken up by the thickened metal in the wall. The drawing load then required to force the thickened metal to flow between the punch and die increases to values several times the maximum load that is required to draw a blank without ironing. For severe cases the deformations required

were beyond the limit of the material and failure occurred in the top wall section. The thickening in a circular cup is nearly uniform at any given level around the cup wall and consequently the ironing failures that occurred in circular cups were similar to that shown in Fig. 3D. The metal flow in a square box causes increased thickening in the region near the center of the side wall just under the point of

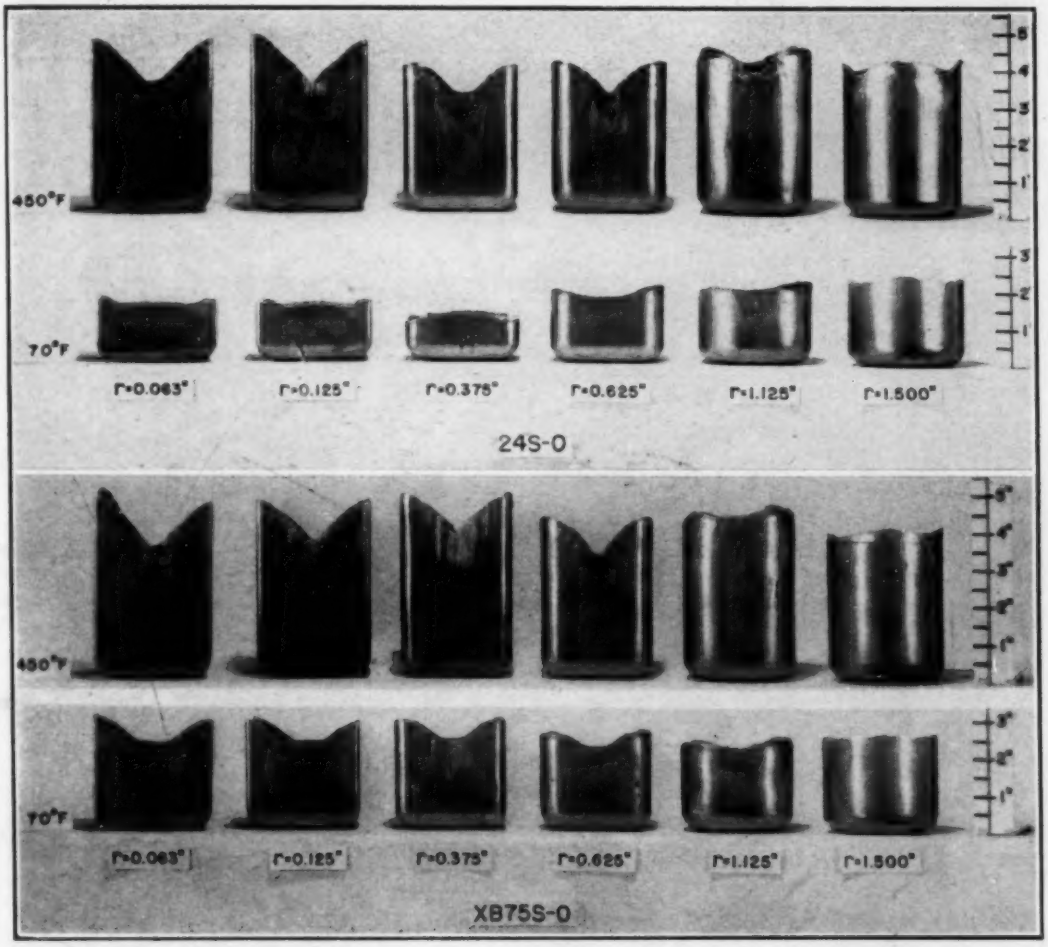

Fig. 5—Maximum Draws at 70 and 450 Degrees Fahr. for Various Corner Radii. 24S-O and XB75S-O aluminum alloys.

minimum height. As a result, ironing tears in square boxes were limited to this region as illustrated in Fig. 3E.

F, G, and H. *Lip Fractures*—The most common type of failure encountered near the critical draw was lip fracture pictured in Figs. 3F, 3G, and 3H. By polishing the edges of the blank it has been found possible to achieve greater draws with some of the T temper alloys that exhibited lip fractures. The fractures illustrated in Figs. 3G and 3H, however, were present in cups drawn from polished

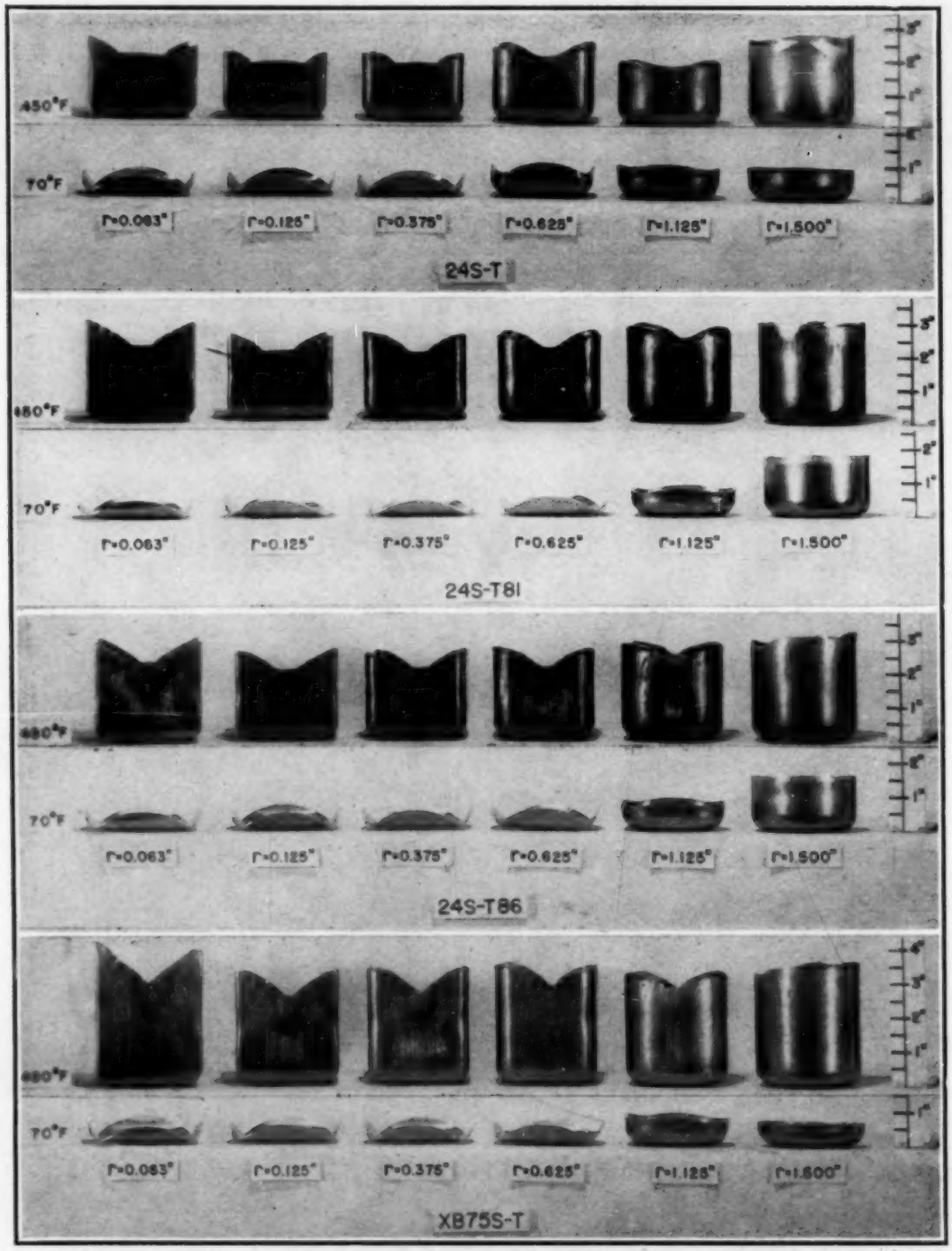

Fig. 6—Maximum Draws at 70 and 450 Degrees Fahr. for Various Corner Radii for Hardened Temper Aluminum Alloys.

edge blanks indicating factors other than the edge condition are involved.

I. *Bend Cracks*—Bend cracks at each corner were encountered when drawing the T temper alloys at 70 degrees Fahr. These fail-

ures were not observed on any test at 450 degrees Fahr. (230 degrees Cent.). Bend cracks are believed to occur when the bend at the intersection of the corner radius and bottom radius was smaller than permissible.

O and W. *Wrinkles*—Although wrinkling of the flange does not constitute the same type of failure as cracking or tearing, it does represent an unsuccessful draw. The wrinkled flange when drawn over

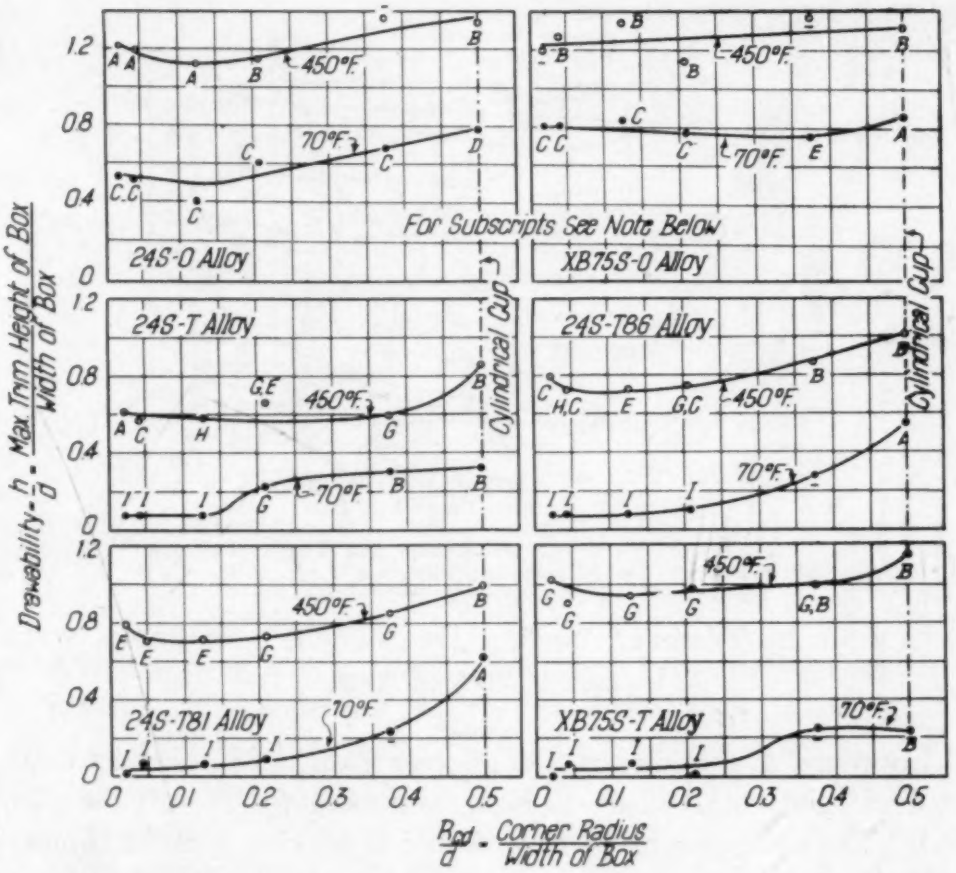

Fig. 7—Effect of Corner Radius and Temperature on the Drawability of Aluminum Alloys. Note: Subscripts A to I indicate failure type which limits the draw. Refer to Fig. 3 for illustrations.

the die radius retains the impressions of the wrinkles as shown in Fig. 4. Classification of boxes with wrinkles into those which are acceptable and those which are not acceptable has been done on the basis of the number and depth of the wrinkles below the maximum trimmed height. The cups shown in Fig. 4 are typical of those two groups. Acceptable cups were defined as those which do not have wrinkles with a depth of more than 0.010 inch.

Influence of Hold-Down Pad Design on Drawability—The purpose of the hold-down pad in the normal deep drawing operation is to prevent the formation of excessive wrinkles. To do this successfully, the shape of hold-down pad opening must conform as nearly as possible to the die contour. If too large an unrestrained area is left between the punch and hold-down pad, wrinkles will form during the last part of the draw. Some difficulties of this nature were encountered when a square hold-down opening with zero corner radius was used with large corner radius dies.

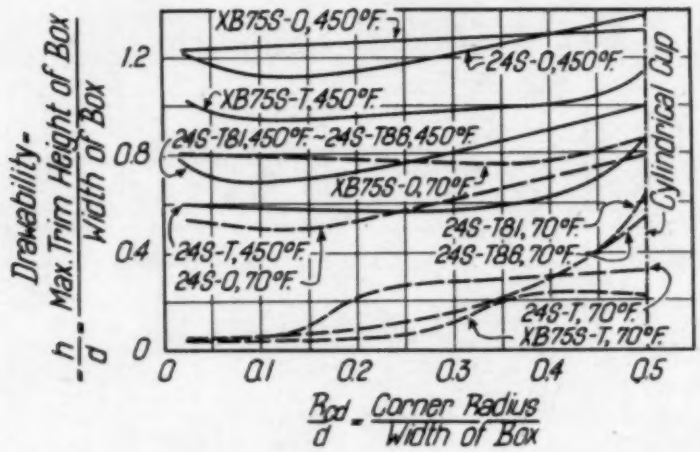

Fig. 8—Effect of Corner Radius and Temperature on the Drawability of Aluminum Alloys.

Influence of Temperature on Drawability of Boxes with Various Corner Radii

Improved drawability at 450 degrees Fahr. (230 degrees Cent.) over drawability at 70 degrees Fahr. was obtained for all the alloys tested. The photographs in Figs. 5 and 6 provide a rapid comparison of the draws obtained for each corner radius at 70 and 450 degrees Fahr. The height of box over the width of box ratios are presented for each alloy in Fig. 7 and are summarized in Fig. 8. The influence of the corner radius on the height of box is small except for the T temper alloys at 70 degrees Fahr. for which the bend crack type of failure limited the drawing. At 450 degrees Fahr. (230 degrees Cent.) bend cracking was not encountered so that great improvement in drawability may be noted in the T temper alloys.

At 450 degrees Fahr. (230 degrees Cent.) all of the T temper alloys tested permitted maximum draws equal to or greater than the maximum draws obtainable from 24S-O alloy at 70 degrees Fahr.

The maximum drawabilities of the T temper alloys tested arranged according to decreasing drawability are in the order of XB75S-T, 24S-T86, 24S-T81, and 24S-T.

The O temper alloys tested, XB75S-O and 24-O, gave a very large increase in drawability at 450 degrees Fahr. (230 degrees Cent.) over that obtained at 70 degrees Fahr.

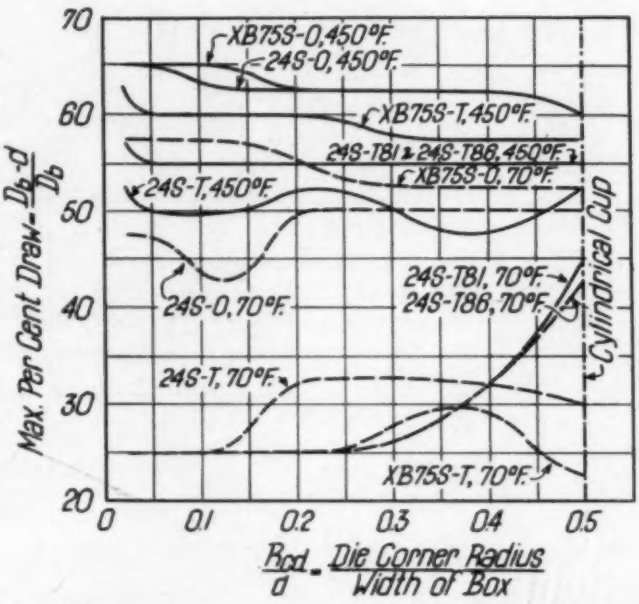

Fig. 9—Effect of Corner Radius and Temperature on the Per Cent Draw of Aluminum Alloys.

The improvements in drawability noted for the few alloys tested are believed to indicate the general trend that may be expected for the elevated temperature deep drawing of the other alloys not investigated.

Also included in Fig. 7 are the symbols representing the type of failure which limited the maximum draw for each condition. These symbols refer to the letters used in Fig. 3 and in the discussion on Types of Failure. In a few cases the evidence was not sufficient to conclude which failure type limited the draw and these points are shown with a dash symbol.

Influence of Temperature on the Per Cent Draw for Boxes with Various Corner Radii

The maximum per cent draw data are given in Fig. 9. The values in Fig. 9 provide a guide for the correct blank size to use in

obtaining the draws reported in the photographs and curves of Figs. 7 and 8.

Influence of Corner Clearance on Drawability

The per cent corner clearance is defined by:

$$\text{Per Cent Corner Clearance} = \frac{cc}{t} \times 100$$

CC = corner clearance
t = sheet thickness
} refer to Fig. 1

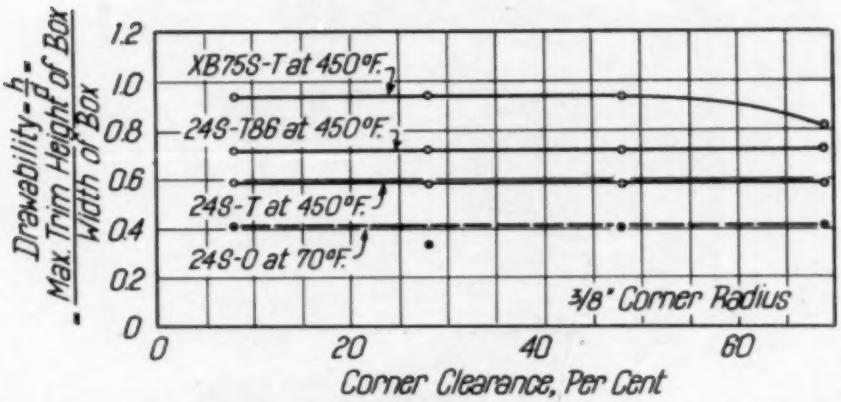

Fig. 10—Effect of Corner Clearance on Drawability of Aluminum Alloys for 3/8-Inch Corner Radius at 70 and 450 Degrees Fahr.

The few tests which were performed to determine the effect of corner clearance indicated that its influence on the drawability was slight over the range of 8 to 69 per cent. Variable corner clearance data for a 3/8-inch corner radius die using XB75S-T, 24S-T86, and 24S-T at 450 degrees Fahr., and 24S-O at 70 degrees Fahr. are presented in Fig. 10.

Influence of Side Clearance on Drawability

The variations in sheet thickness permitted within commercial tolerance may occasionally cause differences in drawability when small side clearances between the punch and die are used. The actual side clearance for the dies used varied between 5 and 12 per cent. The commercial tolerance for 0.064-inch-thick sheet is ± 0.004 inch which is approximately ± 6 per cent. A few tests at 70 degrees Fahr. were made using selected sheet from the same alloy which had the maximum and minimum thickness allowable within

the commercial limits. These tests showed a decrease in drawability for the thicker sheet for which the side clearance was less than 5 per cent.

Use of Developed Blanks

Several developed blanks were investigated in an effort to in-

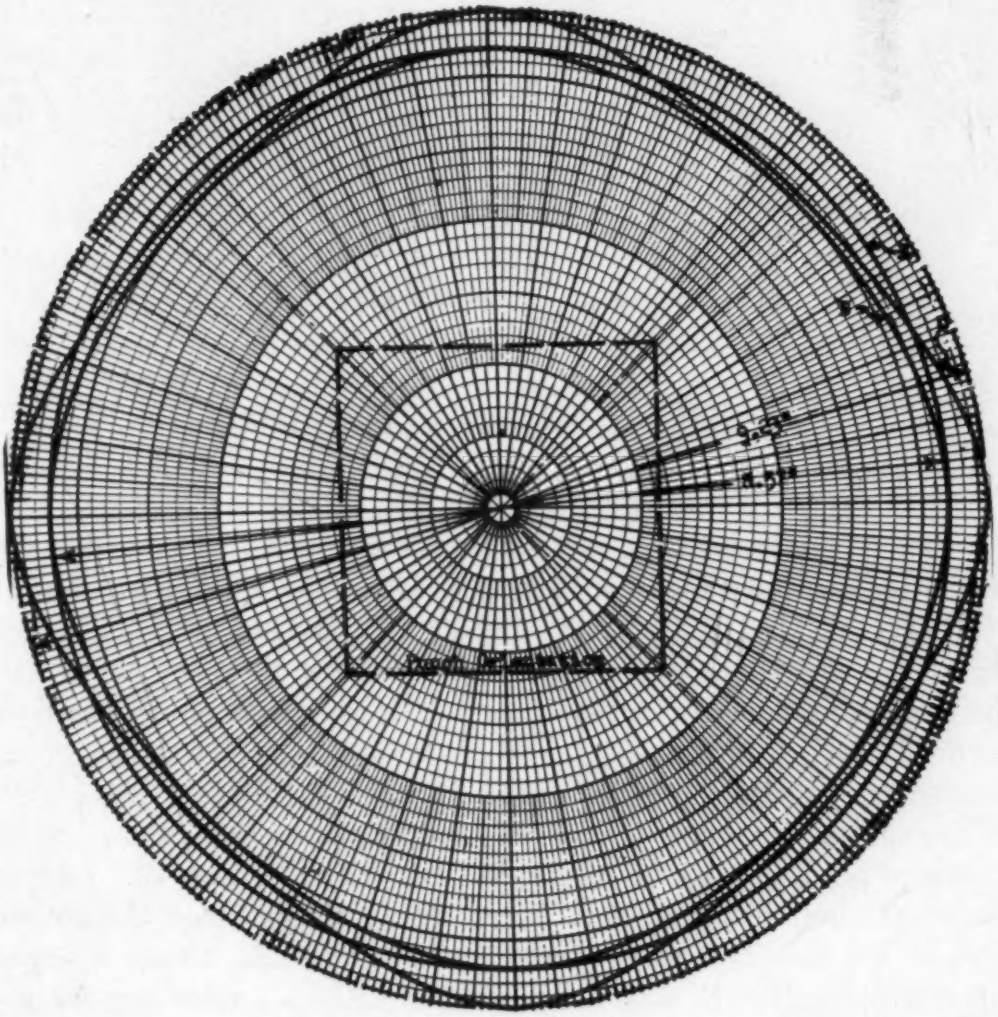

Fig. 11—Blank Development. Scale 0.43.

crease the maximum per cent draw and height of box producing a box without ears. A slight increase in height of box was obtained by removing a small amount of material from the area from which the ears are formed but when sufficient material was removed to produce a box without any ears, severe wrinkles were formed along the corners. In general a circular blank shape was the most satisfactory for producing square box-type parts.

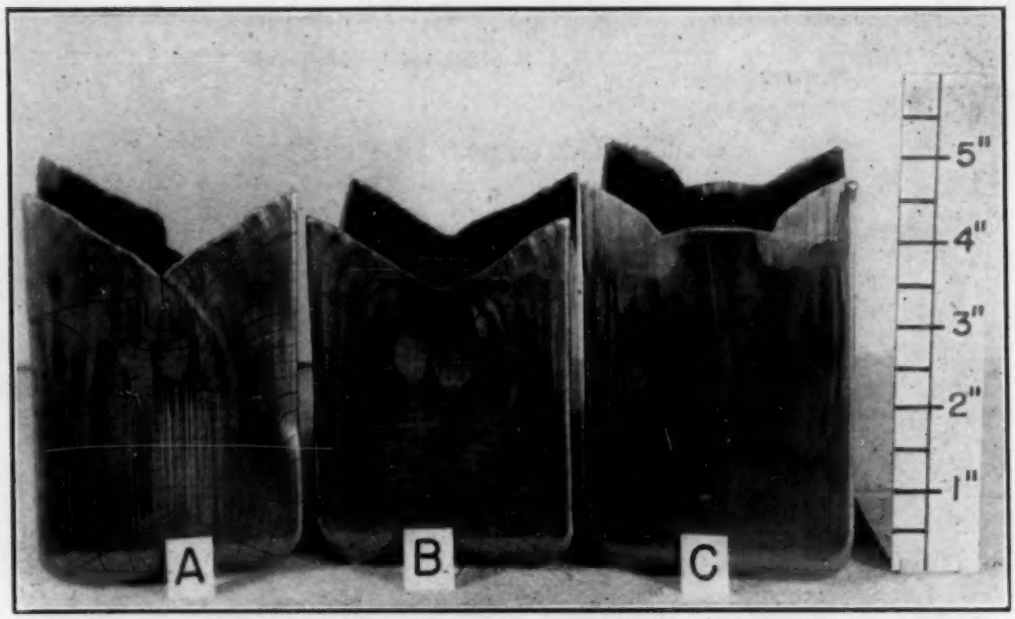

Fig. 12—Cups Drawn from Gridded and Developed Blanks of XB75S-T Alloy. A—8.57-inch diameter gridded round blank. B—8.57-inch maximum dimension developed blank. C—9.23-inch maximum dimension developed blank. Using: Die temperature—450 degrees Fahr. Punch and die corner radii—0.063 inches. Corner clearance—48 per cent.

The blank modifications used are shown in Fig. 11 and the resulting boxes are shown in Fig. 12. Alloy XB75S-T was used at 450 degrees Fahr. (230 degrees Cent.) with a $\frac{1}{16}$-inch corner radius die. Fig. 12A shows a box drawn from one blank size larger than the maximum using a circular blank. The grid lines reveal the flow conditions which produce high corners and a butt joint at the dip in the center of the wall. By removing some of the metal in the region of the ears as in Blank B, Fig. 11, the height of the drawn box is increased and the earing is reduced as shown at B, Fig. 12. Larger blanks of similar shape led to excessive wrinkling along the corner edge of the box as illustrated in Figs. 11c and 12c. Thus a slight increase in cup height was obtained with XB75S-T alloy but the increased tendency for wrinkles to form along the corners limited the blank development.

Conclusions

1. For all the alloys investigated the drawability at 450 degrees Fahr. (230 degrees Cent.) was substantially better than the drawability at 70 degrees Fahr.
2. At 450 degrees Fahr. (230 degrees Cent.) the T temper alloys

drew as well or better than 24S-O at 70 degrees Fahr. for all corner radii.

3. At 450 degrees Fahr. (230 degrees Cent.) the corner radius had little influence on the height of boxes drawn from T temper alloys whereas a corner radius of at least 5/8 inch was required for boxes drawn from T temper alloys at 70 degrees Fahr.
4. This report provides some of the important data for design and tooling of deep drawn box-type parts in aluminum alloys.

Acknowledgments

This report covers part of an investigation sponsored and completely supported by the Office of Production Research and Development of the War Production Board and supervised by the War Metallurgy Committee. The authors wish to acknowledge the guidance of the WMC throughout this investigation and thank the OPRD for authorizing publication of these data which are based on research done under the "restricted" Project NRC-548, and reported in Report Serial No. W-129.

References

1. W. A. Box and William Schroeder, "Effect of Shape on the Formability of Deep-Drawn Sheet-Metal Parts," *Mechanical Engineering,* October 1944.
2. W. V. Crane, "Plastic Working of Metals," 2nd Edition, John Wiley and Sons, Inc., New York, 1939.
3. William Schroeder and Thomas H. Hazlett, "Technical Developments in High-Production Sheet Metal Forming," S.A.E., *Journal,* (Transactions), Vol. 51, No. 5, May 1943.

DISCUSSION

Written Discussion: By William Schroeder, research engineer, Lockheed Aircraft Corp., Burbank, Cal.

The results presented illustrate the beneficial effects that may be obtained by the use of elevated temperatures for drawing box-type parts from the aluminum alloys. As a consequence this paper fills an important gap in the knowledge of deep drawing.

Not specially mentioned in the paper is the fact that it has been found, at 70 degrees Fahr. at least, that the maximum h/d value determined for one size part also applies to other sizes. (A discussion of this principle occurs in the bibliography of Reference 1.) Assuming that this also applies to forming at elevated temperatures, the data represent a general picture of the formability of material for various values of R_{cd}/d and temperatures.

In the case of the soft tempers as 24S-O and 75S-O the question to be settled for production is whether it is more desirable to use a single operation at 450 degrees Fahr. (230 degrees Cent.) (or some intermediate temperature); or two operations at 70 degrees Fahr. For 24S-O where a limiting value for two operations is available, an h/d value of approximately 1.5 can be obtained at 70 degrees Fahr. as compared to 1.38 in one operation at 450 degrees Fahr. (230 degrees Cent.). In the overlapping range of h/d, the hot method has the advantage of eliminating one operation; however, die heating facilities are required. A further disadvantage of hot forming is the increased tendency to scoring of the sides and the difficulty of removing solid lubricants from the drawn part.

Hot forming of some of the high strength tempers, 24S-T for example, may in some cases be undesirable for the reason that short time periods of heating have been found to reduce the corrosion resistance of the material. This has been discussed in more detail by E. H. Burkart under Part I of this paper.

Written Discussion: By O. A. Wheelon, production design engineer, Douglas Aircraft Co., Inc., Santa Monica, Cal.

This paper is an excellent contribution of quantitative information which should help make deep drawing operations a science rather than an art. Of particular interest in the coverage of the new high strength alloys.

In the past most of our thinking on deep drawing operations centered around nonstructural applications such as junction boxes, tanks, etc., employing low strength materials. There seems to be a trend developing where the high strength alloys, particularly 75S, offer considerable advantage and weight saving in primary structural applications such as corner fittings, and secondary structural applications such as seats and berths.

Corner fittings, etc., are correctly made from forgings and impact extrusions where die costs are high and subsequent machining excessive and costly. Drawn boxes are merely trimmed to make several fittings and offer considerable savings. The gages are reasonably heavy and heat treat distortion is not serious so that we could produce either "SO" parts at room temperatures or hot "T".

In the other category gages are lighter and heat treat distortion serious. This would require incomplete drawing, solution heat treat and rehitting to complete in the as-quenched or ice-boxed condition. An alternative would be to draw the fully heat treated and aged material at elevated temperatures.

It appears that the hot operations would represent a production economy, particularly where rehitting and ice-boxing is required. Even more important than the cost differentials is the question of facilities, since most plants shifting to the high strength materials find heat treating facilities overtaxed. This is particularly true of aging.

The question that will decide the issue, however, is which alternative produces the best part. With 75S material strain hardening is deleterious and while heat minimizes this effect we need to make a careful evaluation of the material in the finished part.

The nature and magnitude of residual stresses are important and stress cracking should be checked. The physical properties of the finished part should be carefully checked with particular scrutiny to the compressive yield point and the ductility. The local ductility for both simple tension and bi-axial tension

(stress ratio 2:1) together with the strain hardening characteristics may give an indication of the fatigue characteristics.

Until these points show that equally satisfactory parts can be produced hot from fully aged material we will continue work in the soft condition at room temperature and its attendant processing difficulties. Heat will undoubtedly be used as a means of accomplishing, in a single stage, draws which would otherwise require several stages.

Written Discussion: By F. B. Dahle, supervisor of production research, Battelle Memorial Institute, Columbus, Ohio.

I have followed with considerable interest the authors' progress in their investigation on the "Deep Drawing of Aluminum Alloys at Elevated Temperatures". They have contributed a great deal to our knowledge of deep drawing, and I should like to congratulate them for their painstaking work.

In reviewing this paper, I should like to discuss briefly the test method used in determining the drawability of aluminum alloys. At Battelle, we have felt for some time that the cup drawing test method for determining inherent drawability in a material has considerable merit. Our experience on both brass and steel would indicate that the test, when properly conducted, is capable of showing differences not revealed by conventional tests. The work on aluminum by Dr. Dorn and his co-workers was, therefore, of considerable interest to us in lending confirmation to our results.

On page 295, the authors mention that blanks for the T alloys were prepared with machined edges in order to eliminate the effect of edge conditions. I should like to ask the authors whether any work was done on sheared blanks and, if so, whether any important effects were noted. Sheared edges would, of course, be used on most blanks for commercial drawing operations. In our work on determining the drawability of steel, we could find no important differences between sheared and machined blanks. We did not have an opportunity to determine whether the drawability of brass was affected by the method of edge preparation.

Oral Discussion

B. S. Myers:[3] Elevated temperature forming permits some highly stressed parts to be produced which otherwise could not be formed with certain high strength aluminum alloys. 24S-T86, for example, due to cold work and elevated temperature precipitation treatment, has very high yield strength, very low elongation, and very poor forming characteristics at room temperature. The formability of this alloy is considerably improved when the metal and dies are heated to about 375 or 400 degrees Fahr. (190 or 205 degrees Cent.), thus allowing more numerous applications of this alloy.

The elevated temperature precipitation treatment approaches the hot forming range worked with by Messrs. Finch, Wilson and Dorn. Therefore, in many cases, the preheat operation may be eliminated by forming the metal immediately upon removal from the aging furnace.

I think the work of the authors is certainly commendable, and will prove advantageous in industry.

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[3]Metallurgist, Reynolds Metals Co., Louisville, Ky.

Authors' Reply

The authors are pleased to have the expert comments of Messrs. Schroeder, Wheelon, Dahle and Myers who have been actively engaged in research on forming metals. Although hot deep drawing has some outstanding advantages over the usual practice of drawing at atmospheric temperatures, it also introduces additional production difficulties. The authors believe that the decision for employing hot drawing techniques will require a careful analysis of all of the factors affecting production and the properties of the drawn part. The comments of Messrs. Wheelon and Schroeder are pertinent to this problem for they serve to indicate the present trends of industry toward adoption of high temperatures for deep drawing operations.

The theoretical analyses for determining the maximum draws that may be achieved are yet incomplete. This is particularly true for drawing box-type parts. As far as the authors are aware the drawability of metals cannot yet be inferred from simple test data. Fortunately the investigations of Box and Schroeder (1) have shown that the same per cent draw is obtained for dimensionally similar objects. Thus the possibility of drawing a large part may be ascertained by investigations on small prototype models.

Improvement in drawability by polishing the blank edges was obtained only for a few of the precipitation-hardened alloys that exhibited lip fractures. For such alloys greater draws were achieved when the blank edges were polished. For all alloys that did not reveal lip fracture no improvement in drawability was obtained by polishing the edges.

NEW ALUMINUM ALLOYS, CONTAINING SMALL AMOUNTS OF BERYLLIUM

By R. H. Harrington

Abstract

These new heat treatable compositions are of two types: (a) aluminum-copper-beryllium with the copper and beryllium in the critical ratio of 7 to 1 and (b) aluminum-copper-beryllium-cobalt with the cobalt and beryllium in the critical ratio of 6.5 to 1. These alloys develop superior strength properties as castings made by gravity-sand, centrifugal, and lost-wax methods combined with high thermal stability and unusual oxidation-corrosion resistance. They also develop useful wrought properties so that it is possible for one representative composition to meet requirements in both fields.

Introduction

IN periods of war, time is the essential factor. New alloys, under such conditions, cannot receive complete fundamental testing as for such properties as endurance limit and corrosion resistance. Because of the value of time, new alloys cannot be considered for application simply on the basis of more efficient use when established commercial alloys, already in production and meeting established specifications, are available. Thus the primary war applications for new alloys are those which cannot be adequately met by standardized materials. There have been several of these for the new aluminum alloys described herein and obviously these must remain undisclosed. However, for one result, the field of centrifugal casting for high strength heat treatable aluminum alloys has been greatly widened.

Today it is practically impossible to make a complete search of the literature. Probably for that reason only one reference, pertinent to these new compositions, was found.[1] These investigators described binary alloys of aluminum and beryllium and certain compositions of

[1]R. S. Archer and W. L. Fink, "Aluminum-Beryllium Alloys," *Proceedings*, Institute of Metals Division, American Institute of Mining and Metallurgical Engineers, 1928, p. 616-646.

A paper presented before the Twenty-seventh Annual Convention of the Society held in Cleveland, February 4 to 8, 1946. The author, R. H. Harrington, is research metallurgist, General Electric Research Laboratories, Schenectady, N. Y. Manuscript received July 12, 1945.

aluminum-beryllium-copper, with various ratios of copper to beryllium. All data were for alloys solution-quenched and aged at room temperature, both as cast and for rolled sheet. No complete precipitation heat treatments were disclosed, and the tensile properties were equalled by, or mostly inferior to, those of aluminum alloys of standard commerce. In those days of only 17 years ago, copper-beryllium master alloys were not available and the reported compositions were made with difficulty from metallic beryllium.

Compositions[2]

Compositions of interest to the present research are of two types:

A. (2.0-6.0) copper + (0.1-1.0) beryllium + aluminum balance, specifically with the ratio of copper to beryllium as about 7 to 1; for example, an alloy of 3.5 copper-0.5 beryllium-96 aluminum. Beryllium in excess of about 0.75 per cent tends to cause a decrease in the useful properties of these alloys.

B. (2.0-6.0) copper + (0.1-1.0) beryllium + (0.65-6.5) cobalt + aluminum balance, and specifically with the ratio of cobalt to beryllium as about 6.5 to 1, forming a CoBe phase and leaving the copper free to react largely as in binary aluminum-copper alloys; for examples, such alloys as (1) 3.5 copper-0.5 beryllium-3.2 cobalt-balance aluminum, and (2) 3.8 copper-0.2 beryllium-1.3 cobalt-balance aluminum. Keeping the copper constant, increasing the amount of cobalt-beryllium results in an increase in elastic properties and hardness. However, beryllium in excess of 0.7 per cent and cobalt in excess of 4.5 per cent result in alloys difficult to cast and of inferior properties.

Melting Procedure

A. *Aluminum-Copper-Beryllium Alloys*—Melting and alloying can be done in a standard gas or oil-fired crucible furnace (or in a high frequency furnace for producing ingot). The most economical and apparently best metallurgical method for obtaining a 7 to 1 ratio of copper to beryllium (as for the phase CuBe) is to add a special master alloy of 12.5 beryllium-87.5 copper to the molten aluminum. Beryllium in excess of the 7 to 1 ratio adds nothing to the usefulness of these compositions and is, in fact, damaging to the properties

[2]U. S. Patent No. 2,373,678 and U. S. Patent No. 2,394,546.

in the higher alloy ranges. Beryllium in lesser amounts than by the 7 to 1 ratio can readily be achieved in the alloy most economically by (a), adding to the molten aluminum the desired amounts of standard commercial copper-beryllium master alloys containing from 3 to 5 per cent beryllium or, (b), by addition of the special 12.5 beryllium-87.5 copper master alloy and diluting the beryllium by addition of desired amounts of OFHC (or equivalent purity) copper. A typical melt schedule is as follows:

1. Melt the required charge of 99.85 per cent (or higher purity) aluminum and raise the temperature of the melt to 800 to 850 degrees Cent. (1470 to 1560 degrees Fahr.).

2. Add the required amount of copper-beryllium master alloy, allowing an excess of 0.02 per cent to act as deoxidizer and use *no* flux.

3. Allow the alloyed melt to just freeze in the crucible and then remelt and heat just to 700 to 715 degrees Cent. (1290 to 1320 degrees Fahr.) (for gravity sand castings) and pour immediately. This freezing and remelting largely eliminates the gas (chiefly hydrogen) dissolved in the first melting up to the higher temperature.

B. *Aluminum-Copper-Beryllium-Cobalt Alloys*—The melting procedure is similar to that for the alloys without cobalt. Cobalt is added, either solid (rondelles) or melted, or (preferably) in the form of previously prepared 80 aluminum-20 cobalt master alloy made by dissolving the cobalt in molten aluminum at about 1000 degrees Cent. (1830 degrees Fahr.). This aluminum-cobalt master alloy is easily compounded in a small high frequency furnace. A typical melting schedule is as follows:

1. Melt the required charge of 99.85 (or higher purity) aluminum and raise the temperature of the melt to about 900 degrees Cent. (1650 degrees Fahr.).

2. The 80 aluminum-20 cobalt master alloy is added, while the melt is quietly stirred with a graphite rod.

3. The melt is cooled to about 800 degrees Cent. (1470 degrees Fahr.) and the desired amount of copper-beryllium master alloy is added.

4. As soon as the copper-beryllium is dissolved, the melt is allowed to just freeze in the crucible.

5. As soon as solidification is complete, the alloy is remelted and poured immediately in the temperature range of 710 to 725 degrees Cent. (1310 to 1335 degrees Fahr.). When the cobalt content ex-

ceeds about 3 per cent, there may be difficulty from segregation, and quiet stirring with a graphite rod is advised.

Four Compositions, Representative of Alloys Yielding High Elastic Properties and Hardness

These four alloys are chosen because (a) they represent economical and efficient combinations of useful properties, (b) they have been subjected to, and justified by, practical conditions in the foundry, etc., and (c) they serve for purposes of comparison.

No. 1. 3.8 Cu–0.2 Be–96 Al (Cu:Be as 19:1)
No. 2. 3.5 Cu–0.5 Be–96 Al (Cu:Be as 7:1)
No. 3. 3.8 Cu–0.2 Be–1.3 Co–94.7 Al (Cu:Be:Co as 19:1:6.5)
No. 4. 3.5 Cu–0.5 Be–3.2 Co–92.8 Al (Cu:Be:Co as 7:1:6.5)

Compositions 1 and 2 are quoted to indicate the increase in elastic properties, etc., due to a ratio of 7 to 1 for copper to beryllium (CuBe compound atomic weight ratio) as in No. 2.

Compositions 3 and 4 are given to show the effect on tensile properties, etc., of an increase in CoBe relative to the copper content (atomic weight ratio of 6.5 to 1 for cobalt to beryllium). In these alloys with cobalt, the ratio of copper to beryllium is not critical. It is, however, the ratio of 6.5 to 1 for cobalt to beryllium that *is* critical (with copper present, preferably, from 3.5 to 4.5 per cent). Alloys of aluminum-beryllium-cobalt, without copper, are only slightly precipitation-hardenable and very inferior as to properties.

Heat Treatments

All four compositions are strongly precipitation-hardenable. Thorough testing of tensile bars has shown that these alloys, in the cast and heat treated condition, develop maximum strength with maximum hardness. Each alloy was subjected to the same preliminary heat treatment study. Half-inch-thick disks were cut from 1-inch diameter cast rods. Individual samples were then quenched from 4 hours solution treatment at temperatures of 510, 520, 530, 540, 550, 560 and 570 degrees Cent. (950, 970, 985, 1005, 1020, 1040 and 1060 degrees Fahr.) (except that alloy No. 2 began to melt at 570 degrees Cent.). Individual samples, so quenched from each solution temperature, were then aged 4 hours at each of the following reheat temperatures: 50, 75, 100, 125, 150, 175, 200, 225, 250, 275 and

300 degrees Cent. (120, 165, 255, 300, 345, 390, 435, 480, 525 and 570 degrees Fahr.). Hardness values for each treatment (Rockwell B scale) thus indicated the most effective solution and aging temperatures. Solution temperatures of 530 to 550 degrees Cent. (985 to 1020 degrees Fahr.) and aging temperatures of 100 to 225 degrees Cent. (210 to 435 degrees Fahr.) develop useful properties in these alloys. Solution temperatures over 550 degrees Cent. (1020 degrees Fahr.) cause grain boundary melting and spoiling of the material.

The effect of time at solution and aging temperatures was then studied, using periods of 2, 4, 8 and 20 hours in each case. From these data were chosen the preferred complete heat treatments to maximum hardness, bearing in mind also relative economies such as the 8-hour work day, etc. Thus a sacrifice of 1 to 3 points on the Rockwell B scale can satisfactorily be made for the sake of production economy. The preferred (or critical) precipitation hardening treatments (for castings) were determined to be as follows:

Alloy No.	Composition	Preferred Heat Treatment	Rockwell B
1	3.8 Cu–0.2 Be–Al	4 hrs., 540° C., W.Q., 8 hrs., 200° C.	45–48
2	3.5 Cu–0.5 Be–Al	4 hrs., 540° C., W.Q., 8 hrs., 200° C.	55–57
3	3.8 Cu–0.2 Be–1.3 Co–Al	4 hrs., 540° C., W.Q., 6 hrs., 200° C.	58–62
4	3.5 Cu–0.5 Be–3.2 Co–Al	4 hrs., 540° C., W.Q., 2 hrs., 200° C.	71–75

W.Q. = water quench. Solution temperature control = ±5 degrees Cent.

It was established that, for alloys of aluminum-copper-beryllium, superior hardness is developed by employing the critical atomic weight ratio of 7 to 1 for copper to beryllium as in the phase CuBe. It was also established that the hardness is further improved by the addition of cobalt to such compositions in the critical atomic weight ratio of 6.5 to 1 for cobalt to beryllium as in the compound CoBe.

Tensile Properties of Sand Cast Alloys

These tensile properties were obtained from sand cast A.S.T.M. standard (0.505-inch) tensile bars after applying the preferred complete precipitation hardening heat treatment for each alloy (for the new compositions, as noted under the section on heat treatment). Table I also gives the comparison with two well-known sand casting alloys of commerce.

The effect of heat treatment on the tensile properties of the new aluminum-copper-beryllium ± cobalt alloys is quite critical but well within present-day heat treatment practice (±5 degree Cent. control on the solution temperature). For brevity, Table II shows the effects

Table I
Standard Tensile Properties for H-T Sand Casting Alloys

Alloy No.	Composition (Bal. is Al) Per Cent	Proportional Limit	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.	Hardness Rock. B
195	4 Cu	15,000	22,000	36,000	5	40
356	7 Si–0.3 Mg	13,000	22,000	32,000	3	30
356H	Special*	19,000	30,000	38,000	3	42
1	3.8 Cu–0.2 Be	22,000	32,000	41,000	4	47
2	3.5 Cu–0.5 Be	25,000	35,000	44,000	5	55
3	3.8 Cu–0.2 Be–1.3 Co	25,000	34,000	43,000	3	62
4	3.5 Cu–0.5 Be–3.2 Co	32,000	40,000	41,500	1	72

*Critical heat treatment: U. S. Patent 2,246,134.

of various heat treatments on Aluminum-copper-beryllium 3 as representative of this class of compositions.

Natural Aging, Straightening and Mild Cold Shaping of Castings

"Natural aging" consists simply of allowing the solution-quenched alloys to age at room temperatures. Some precipitation

Table II
Effect of Various Heat Treatments on the Tensile Properties of Aluminum-Copper-Beryllium 3. (3.8 Copper-0.2 Beryllium-1.3 Cobalt)

	Heat Treatment							
	Solution		Aging					
No.	Time Hrs.	Degrees Cent.	Time	Degrees Cent.	Proportional Limit	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.
1	.	...	As Cast	...	8,200	17,000	22,200	2.5
2	4	540	Natural Aging		14,000	25,000	30,100	1.5
3	4	540	8 hrs.	150	10,900	22,200	31,250	4
4	4	540	20 hrs.	150	14,250	24,050	33,100	4
5	4	540	8 hrs.	175	15,000	24,300	35,000	3.5
6	4	540	20 hrs.	175	19,000	28,000	37,000	3
7	4	540	8 hrs.	200	25,000	34,000	43,500	3
8	4	540	20 hrs.	200	26,000	35,000	44,200	3
9	4	530	20 hrs.	150	12,500	20,000	30,600	3.5
10	4	530	20 hrs.	175	14,000	25,000	35,000	3
11	4	520	20 hrs.	150	6,000	15,100	25,050	3.5
12	4	520	20 hrs.	175	7,000	19,200	28,950	3

No. 1 As-Cast. Naturally Aged 6 weeks.
No. 2 Solution Treated. Naturally Aged 6 weeks.

hardening alloys will actually effect precipitation with the passing of time at ordinary temperatures while others require elevated temperatures to initiate and develop the reaction. Generally, with aluminum alloys, natural aging does not develop maximum properties and "artificial aging" (at elevated temperatures) is required to develop consistently reproducible properties for engineering applications.

Individual samples of all four new compositions were solution quenched respectively from 4-hour heats at 510, 530, 540, 550 and 560 degrees Cent. (950, 985, 1005, 1020 and 1040 degrees Fahr.), and hardness was measured with the passing of time at room temperature. All 4 new alloys naturally age very rapidly with maximum effects from the 540 degree Cent. (1005 degree Fahr.) solution treatment.

Table III
Natural Age Hardening Effects (Rockwell B Hardness). 540 Degree Cent. Solution Treatment

Time	Alloy 1	Alloy 2	Alloy 3	Alloy 4
5 minutes	–1B	27B	27B	37B
1 hour	12	44	42	40
3 hours	25	48	45	43
6 hours	29	48	44	47
24 hours	32	50	53	48
2 days	37	50	53	53
5 days	37	50	53	55
6 days	37	50	52	58
7 days	37	50	52	57
8 days	37	50	52	58
9 days	36	49	52	59
12 days	37	49	52	60
19 days	38	51	53	61
4 weeks	38	51	54	62
Fully Treated	47	55	62	72

Four weeks natural aging practically completes this effect. Comparison of the 4-week natural aging hardness values with those for maximum artificial aging (Fully Treated, Table III) shows that natural aging achieves about 90 per cent of the full hardening.

Thus it is seen that these alloys naturally age very rapidly and to a marked degree. The chief value of the natural age hardening data is to show that, if these castings are to be cold-worked after the solution treatment, such cold working must be done within 1 day after quenching for alloy No. 1 and preferably within 1 hour after quenching for alloys 2, 3 and 4.

To demonstrate the cold workability of the solution-quenched cast materials, 1-inch diameter solution-quenched cast rods can be cold-swaged the following amounts:

Alloy No. 1.....50 Per Cent
Alloy No. 2.....50 Per Cent
Alloy No. 3.....23 Per Cent
Alloy No. 4.....10 Per Cent

Table IV gives the resultant properties after subsequent artificial aging as for castings.

Table IV
Properties of Cast Materials, Cold-Worked Between Quench and Aging

Alloy	Proportional Limit	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.
1. 3.8 Cu–0.2 Be	30,000	41,000	52,000	10
2. 3.5 Cu–0.5 Be	33,000	43,000	46,000	4
3. 3.8 Cu–0.2 Be–1.3 Co	33,000	42,000	43,000	3
4. 3.5 Cu–0.5 Be–3.2 Co	33,000	40,000	42,000	1

Thermal Stability

The standard sand casting aluminum alloys (such as 195 and 356) are aged at 150 to 160 degrees Cent. (300 to 320 degrees Fahr.) for maximum properties whereas the aluminum-copper-beryllium sand castings are aged at 200 degrees Cent. (390 degrees Fahr.). Table V shows the overaging required to lower the hardness of these new alloys into the range of hardness for alloys 195 and 356, fully heat treated.

Table V
Comparative Thermal Stability of H-T Sand Cast Alloys

Alloy No.	Aging Temperature	Rockwell B Hardness	Degrees Cent. Superiority for Al-Cu-Be
195	160° C.	40	..
356	155° C.	30–42	..
	(over-aging:)		
1	6 hrs., 225° C.	37	65
2	6 hrs., 250° C.	39	90
3	6 hrs., 275° C.	37	115
4	6 hrs., 300° C.	35	140

Thus it appears qualitatively that, compared with the established alloys, the new aluminum-copper-beryllium materials will (a) have higher service temperatures, (b) carry the same stress for the same time at higher temperatures, (c) bear higher stresses for the same life at the same temperature or (d) carry the same stress for longer time at the same service temperature.

Outstanding Oxidation Resistance

Most of the commercial aluminum alloys, when heated to solution temperatures, develop oxide coatings that range from a very dull "pearl" gray to the dark gray for alloys containing copper. A very striking feature of the new aluminum-copper-beryllium composition

is the high luster and apparently perfect retention of the originally bright surface (as cast, machined, or polished) after such solution treatments as 4 to 20 hours at 540 degrees Cent. (1005 degrees Fahr.) followed by a water, or 10 per cent salt solution, quench. It is impossible to tell for certain, by visual examination, and with the two pieces side by side, as to which one has been heat treated. This is in spite of the fact that the aluminum-copper-beryllium alloys contain from 3.5 to 4 per cent of copper. This impressive effect, plus qualitative tests of 4 years of exposure to Schenectady industrial atmosphere, seems to indicate that these new alloys possess high resistance to thermal-oxidation and oxidation-corrosion. Standard corrosion tests, which, after all, give sharply limited information, have not been made due (a) to the volume of drastically needed tests that take precedence and (b) to the frequent change of local testing conditions.

Over-All Rating for Sand Casting

As a result of foundry experience with several parts requiring high strength and of different casting design (inclusive of such features as fluidity, macrostructure, shrink, minimum interdendritic porosity, fine grain size, microstructure, etc.) a strong preference for Aluminum-copper-beryllium 3 has developed, with Aluminum-copper-beryllium 2 a fair second. Aluminum-copper-beryllium 1 would be regarded as very good if the other two did not exist. (Aluminum-copper-beryllium 4 has too low an elongation for consideration as a regular casting alloy although it might well serve where wear resistance is of prime importance such as an occasional lightweight bearing application.)

Centrifugal Castings

Recently a part of difficult design had to be produced in aluminum alloy. It could not be produced by machining wrought stock. It was inherently a design that could not be satisfactorily gravity sand cast as was proven by many attempts on the part of several sources. The past had yielded a present authoritative conception that high strength castings of heat treatable aluminum alloys could not be produced in good quality by centrifugal casting. However, it has been proven that the new aluminum-copper-beryllium compositions yield excellent centrifugal castings.

The following alloys were tested in the form of the actual casting, made according to proven centrifugal practice, developed by substantial tests in casting these alloys:

Alloy No.	Composition
25S	4.5 Cu–0.8 Si–0.8 Mn–balance Al
356H	7 Si–0.3 Mg–balance Al
40E	5 Zn–0.5 Mg–0.6 Cr–0.2 Ti–balance Al
1	3.8 Cu–0.2 Be–balance Al
2	3.5 Cu–0.5 Be–balance Al
3	3.8 Cu–0.2 Be–1.3 Co–balance Al

These castings were made in a metal mold with sand cores. Each alloy was given its specific heat treatment (356, special, U. S. Patent No. 2,246,134). Cross-section macrostructure results as follows:

25S. General, very coarse gas porosity. Insufficient fluidity. Medium grain size.

40E. General, fine gas porosity. Short shrink cracks in critical sections. Grain size: coarse to medium.

356H. Very coarse grain size. Some gas porosity. Rather coarse interdendritic shrinkage. Fair surfaces.

2. Grain size same as for 356 with much less interdendritic shrinkage. More sound than gravity sand casting. Good surfaces.

1. Uniform medium grain. Absolutely no visible shrinkage. Fair surfaces.

3. Uniform very fine grain. Absolutely no visible shrinkage. Excellent surfaces.

Few gravity sand castings reproduce tensile properties equal to the properties obtained from the ideally cast tensile bars. Tensile bars were machined from actual castings made by both methods: (a) gravity sand casting and (b) centrifugal casting (exception: no gravity casting of alloy 25S). The bars were then heat treated as recommended for each alloy. Table VI compares the properties obtained for the two types of castings for each alloy except that, for alloy 25S, typical wrought properties are compared with those of the centrifugal casting.

Table VI

Properties of Gravity Sand Castings Versus Centrifugal Castings

Alloy	Type of Casting	Proportional Limit	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.
25S	Wrought	.....	35,000	57,000	18.0
	Centrifugal	.....	39,500	47,000	1.2
40E	Gravity	13,000	22,000	30,200	5.0
	Centrifugal	15,300	24,400	26,600	5.2
356H	Gravity	18,000	27,000	34,200	2.5
	Centrifugal	19,800	29,600	32,200	3.7
2	Gravity	22,000	31,000	39,100	4.5
	Centrifugal	15,200	23,750	35,500	11.0
1	Gravity	20,000	28,000	36,200	3.6
	Centrifugal	23,750	31,200	36,750	5.8
3	Gravity	22,500	30,600	38,100	2.7
	Centrifugal	26,700	38,000	38,600	3.0

Aluminum-copper-beryllium 3 was outstanding in macrostructure, properties and performance.

Lost Wax Casting

It became necessary to produce another part in aluminum alloy by the method of lost wax casting. The part was of such a shape as to make it impossible for tensile bars to be machined from it. Hence choice of the most suitable alloy had to be based upon the strength properties of lost wax cast standard tensile bars. The very slow cooling, inherent to this method of casting, could be expected to be detrimental to aluminum alloys in general.

On account of its all-around good casting properties Aluminum-copper-beryllium 3 was chosen from the new compositions for comparison with alloys 40E, 195 and 356H. The average results from tests of 12 tensile bars are given in Table VII. The bars of each alloy were given the required heat treatments (356H: special treatment for high strength, U. S. Patent No. 2,246,134).

Table VII
Lost Wax Cast Tensile Bar Properties

Alloy	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.
40E	23,500	28,800	3
195	23,750	30,760	3
356H	30,270	35,775	3
3	35,200	44,100	3

Again the new alloy, with all of its favorable properties, appears to be outstanding.

Forging

Unfortunately, due to circumstantial conditions (war) previously noted, forging tests on only one of the new compositions, Aluminum-copper-beryllium 3, are available at present.

It has been previously noted that Aluminum-copper-beryllium 1, 2 and 3 can be substantially cold-worked after solution treating the cast stocks. It is also possible to substantially cold work these three compositions after they have been annealed above 350 degrees Cent. (660 degrees Fahr.). Introductory tests also show that these new alloys can be suitably hot-forged by orthodox methods in the temperature range of 500 down to 350 degrees Cent. (930 to 660 degrees Fahr.).

A portion of a billet of Aluminum-copper-beryllium 3, 5 by 4½ by 4½ inches, was soaked 5 hours at 455 degrees Cent. (850 degrees Fahr.) and triple upset to final dimensions under a 2500-pound ham-

mer. Tensile bars were machined from this stock. The 4 bars were water-quenched from 4 hours solution at 540 degrees Cent. (1005 degrees Fahr.). Two bars were then aged for 16 hours at 155 degrees Cent. (310 degrees Fahr.) and the other two were aged for 6 hours at 200 degrees Cent. (390 degrees Fahr.). Tensile properties resulted as follows:

Aging Temp.	0.2 Per Cent Yield Strength	Tensile Strength	Per Cent Elong.
155° C. (16 hrs.)	37,200	53,750	12
200° C. (6 hrs.)	37,800	52,600	4.5

The 155 degree Cent. (310 degree Fahr.) aging treatment may be preferable for most forging applications due to the higher resulting elongations while the 200 degree Cent. (390 degree Fahr.) aging treatment indicates unusual thermal stability.

A part of Aluminum-copper-beryllium 3, so forged, machined and cold-shaped after solution treatment, and aged for 16 hours at 155 degrees Cent. (310 degrees Fahr.), gave a performance test slightly better than an orthodox forging of wrought aluminum alloy 25ST so that it might be safely said that these two alloys were equivalent on a strength basis.

When time and conditions permit, forging tests will be made on Aluminum-copper beryllium alloys 1 and 2 as it is not unlikely that they might be somewhat superior to Aluminum-copper-beryllium 3 for forgings.

Conclusions

The general conclusions are as follows:

1. The new alloys of aluminum-copper-beryllium, with and without cobalt, are outstanding for their unusually high tensile properties combined with high thermal stability and oxidation resistance.

2. These new compositions have proven themselves in practice by the following methods of casting: Gravity sand casting, lost wax method of precision casting, and centrifugal casting.

3. The result of foundry experience, so far, with these new alloys, tends to favor Aluminum-copper-beryllium 3 (3.8 copper-0.2 beryllium-1.3 cobalt-balance aluminum) for its all-around casting properties.

4. Comparison of Aluminum-copper-beryllium 1 with alloy 195, Table I, shows the marked beneficial effect of small additions of beryllium to aluminum-copper alloys.

5. Comparison of Aluminum-copper-beryllium 2 with Aluminum-copper-beryllium 1, Table I, is indicative of the benefit from the critical ratio of copper to beryllium of 7 to 1, as for the phase CuBe.

6. Aluminum-copper-beryllium 3, compared with Aluminum-copper-beryllium 1 in Table I, is an example of the benefit to tensile properties to be gained by adding cobalt to aluminum-copper-beryllium alloys in the ratio of 6.5 to 1 for cobalt to beryllium (as in the phase CoBe). Such an addition of cobalt to aluminum-copper-beryllium alloys results in two basic phases ($CuAl_2$ and CoBe types) for precipitation hardening treatment.

7. Below the noted maximum alloy contents, increase in the amount of the CoBe-type constituent effects increase in the "elastic properties" (proportional limit and 0.2 per cent yield strength) and hardness as indicated by comparing Aluminum-copper-beryllium 4 with Aluminum-copper-beryllium 3 in Table I.

8. There is good evidence that the wrought forms of such alloys as Aluminum-copper-beryllium 1, 2 and 3 combine useful tensile properties with high thermal stability and high oxidation resistance.

Acknowledgments

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DISCUSSION

Written Discussion: By W. W. Edens, technical director, Ampco Metal, Inc., Milwaukee.

The author is to be congratulated, not only for the excellence of the paper, but for contributing to the engineering world another family of alloys, this time in the light metal field. His previous work with the copper-beryllium-chromium and the copper-beryllium-cobalt alloys which are now playing such an important part in the resistance welding industry and his work on the role of strain in precipitation reactions in alloys are well known.

It was the writer's privilege to work in close co-operation with the author on the centrifugal casting phases of this work. The casting referred to by the author was of such section and intricacy as to put a premium on the "castability" of the alloy. Attempts at sand casting were unsuccessful even with these new alloys. There was, of course, a problem of establishing proper thermal gradients as related to foundry practice and progressive solidification; however, only a limited variation could be applied because of the design of the casting leaving a distinct alloy requirement of good "castability". The author's comments on macro-structure and tensile properties of centrifugal casting need no further elaboration.

The alloys were melted in a conventional furnace from pre-alloyed ingot. No flux was used and, in general, the melting practice conformed to what might be considered good practice for any of the aluminum casting alloys with proper emphasis on melting and pouring temperature control. No "special" practices were used and I feel sure that the alloys can be readily handled by any foundry possessing reasonable metallurgical and foundry control.

The high physical properties of these alloys, with special recognition to their excellent proportional limits, mark them to be worthy of engineering recognition and of real potential value; however, the writer feels that the less explored resistance to oxidation as reported, and the corrosion resistance this implies, may prove to be an outstanding virtue in these alloys. It is to be hoped that the author will have the opportunity of fully exploring the corrosion resistance of these alloys, reporting them at a later date as a further contribution to light metal alloy metallurgy and engineering.

Written Discussion: By C. B. Sawyer, president, Brush Beryllium Company, Cleveland.

Those interested, like myself, in the promotion of uses for beryllium are sure to be pleased with the encouraging results here reported in detail on small beryllium additions to strong aluminum alloys.

Dr. Harrington's composition No. 1 consisting of 3.8 per cent copper, 0.2 per cent beryllium with balance largely aluminum is suggestive of a German casting alloy known as "Hydronalium No. 511" consisting of 5 per cent magnesium, 1 per cent silicon, 0.2 per cent beryllium, 0.3 per cent manganese, 0.15 per cent titanium, balance largely aluminum, wherein magnesium replaces the copper of Dr. Harrington's alloy. The German alloy came to the attention of the writer and accompanying British representative, H. A. Sloman, last summer when interviewing Drs. Schultz and Siebel of the I. G. Farbenindustrie of Bitterfeld when they were residing near Frankfort in the American Zone of Occupation.

The resulting Beryllium Industry report to the U. S. Government is now declassified for reproduction and distribution by the Bureau of Mines of the Department of the Interior.

According to the Germans cited, the hydronalium alloy is of good thermal conductivity and was used for making sand cast cylinder heads for air-cooled airplane motors. Drs. Schultz and Siebel estimated that about 5000 tons of this alloy was used in this way during the war period, but 2000 tons may be nearer the true figure. The presence of the 0.2 per cent of beryllium kept this casting alloy relatively free from dross and made it possible for it to fill the fine details of the sand mold, including especially the air cooling fins with their small cross sections and large surfaces. For this application it was important to avoid cold shuts and imperfections of casting which would interfere with thermal conductivity and the beryllium addition was notably successful in accomplishing this aim.

Although the primary purpose of the hydronalium alloy was to replace the "nickel alloy", probably Y-alloy, of 4 per cent copper, 2 per cent nickel and 1.5 per cent magnesium, balance aluminum, Drs. Schultz and Siebel expect a permanent use of the principle of small beryllium additions for inhibiting dross formations and conferring liquidity. This same effect is now well known for magnesium base alloys.

Does Dr. Harrington consider this effect on the liquid metal to be significant in his aluminum alloys which contain no magnesium? His observation of resistance to oxidation, when his alloys are heated but still in the solid state, appears to be general for alloys containing beryllium, although the amount of beryllium required to produce the effect varies from one alloy to another.

Written Discussion: By W. Bonsack, director of laboratories, The National Smelting Co., Cleveland.

Dr. Harrington has brought forth a very interesting series of new alloys. He has added a few new precipitation hardening phases to the old Mg_2Si, $MgZn_2$ and the newer ternary aluminum-magnesium-zirconium phases.

Apparently the proportioning of copper and beryllium and copper-cobalt-beryllium is quite critical as in many of the more modern alloys. Since the alloy requires 99.85 or higher purity aluminum, it must be concluded that other alloying elements tend to interfere. It would be interesting to know whether the author has information on the effect of the various common alloying elements such as iron, silicon, zinc, manganese, and magnesium on the properties of these alloys.

In regard to the high purity required for these alloys, their price must be much higher, in some even twice the cost of most common high strength aluminum alloys.

Prevention of contamination must make melting of the alloy and handling of scrap a delicate matter.

The resistance to oxidation of aluminum alloy carrying beryllium has been recognized some time ago, also its beneficial influence on corrosion resistance and it has been utilized particularly in aluminum-magnesium type alloys which are subject to rapid oxidation. Therefore the retention of brightness and luster through heat treatment can be expected on alloys of such high beryllium content.

The physical properties seem to be on a higher level than the commercial aluminum alloys. There are alloys today which give even higher properties with-

out resorting to such high priced material. For example, we have developed alloys of the aluminum-magnesium-zinc type which I will use for comparison.

Alloy	Heat Treatment	Proportional Limit 0.01%	Yield Strength 0.2%	Tensile Strength Psi.	Per Cent Elongation
	SAND CAST				
No. 1	4 Hrs. 540°C—8 Hrs. 200°C	22000	32000	41000	4
No. 2	4 Hrs. 540°C—8 Hrs. 200°C	25000	35000	44000	5
No. 3	4 Hrs. 540°C—6 Hrs. 200°C	25000	34000	43000	3
No. 4	4 Hrs. 540°C—2 Hrs. 200°C	32000	40000	41500	1
Ternalloy 8	6 Hrs. 540°C—8 Hrs. 200°C	31600	37900	41000	1.6
	6 Hrs. 540°C—4 Hrs. 200°C	34900	40500	42500	1.2
	6 Hrs. 540°C—2 Hrs. 200°C	35100	45700	47000	1.0
Ternalloy 7	6 Hrs. 540°C—8 Hrs. 200°C	28300	30500	41600	3.0
	6 Hrs. 540°C—4 Hrs. 200°C	31200	38200	42800	2.0
	6 Hrs. 540°C—2 Hrs. 200°C	28800	36000	41400	2.0
	CHILL CAST				
Ternalloy 8	6 Hrs. 540°C—8 Hrs. 200°C	36100	43500	54300	5.2
	6 Hrs. 540°C—4 Hrs. 200°C	38400	45200	56000	5.4
	6 Hrs. 540°C—2 Hrs. 200°C	38000	45700	55700	4.8
Ternalloy 7	6 Hrs. 540°C—8 Hrs. 200°C	29800	35600	48200	5.8
	6 Hrs. 540°C—4 Hrs. 200°C	28000	34200	47800	7.3
	6 Hrs. 540°C—2 Hrs. 200°C	33400	40400	52000	6.1
Alloy 3 Forged	4 Hrs. 540°C—6 Hrs. 200°C		37800	52600	4.5

There is one set of figures which puzzles me. Table VI, last line, shows 38000 yield strength, 38600 tensile strength, and 3 per cent elongation. I have never seen yield and tensile strength so close and still have 3 per cent elongation. May this be a typographical error?

The alloys are undoubtedly more resistant to higher temperatures as indicated by the fact that high aging temperatures must be employed to develop maximum properties. Immovability of elongation is remarkable. The alloys should be good in high temperature applications.

The author should be congratulated for a fine piece of work.

Written Discussion: By L. W. Kempf, assistant director of research, Aluminum Research Laboratories, Aluminum Company of America, Cleveland.

Dr. Harrington's work again calls attention to the relatively high mechanical properties obtainable in test bar castings of high purity aluminum alloys containing copper as the principal alloying ingredient. The data on page 10, in the A.S.M. publication (1940) AGE HARDENING OF METALS, indicate that sand cast test bars of a high purity aluminum alloy containing about 3.8 per cent copper is capable of giving tensile strengths of the order of 42,000 pounds per square inch with about 10 per cent elongation. The same data indicate that alloys containing about 5.5 per cent copper are capable of developing tensile strengths of the order of 47,000 pounds per square inch with elongations of about 5 per cent. These properties are of the same magnitude as those recorded in Dr. Harrington's paper for aluminum-copper-beryllium-cobalt alloys.

After Dr. Harrington's work was reported, we made up two heats of his alloy No. 3 following the instructions as nearly as possible as given in his paper with respect to the details of alloying, melting, and heat treatment. For

the purposes of comparison, we also made a heat of binary 3.8 per cent copper alloy utilizing aluminum of the same purity as that used for the aluminum-copper-beryllium-cobalt alloy. The results of our tests on separately cast sand cast test bars are presented in the accompanying table. It will be noted that the binary

Mechanical Properties of Sand Cast Tensile Specimens of Al-Cu and Al-Cu-Be-Co Alloys

	Per Cent					Heat Treatment	Yield Strength Psi.	Tensile Strength Psi.	Elong. % in 2 In.
	Cu	Fe	Si	Be	Co				
C2555	3.83	0.06	0.04	0.29	0.95	4 hrs. 540°C, B.W.Q.* 6 wks. r.t.	28,000	42,660	5.3
C2568	3.80	0.12	0.07	0.26	1.30	4 hrs. 540°C, B.W.Q.* 6 wks. r.t.	28,100	41,000	4.0
C2556	3.79	0.10	0.08	....	....	4 hrs. 540°C, B.W.Q.* 6 wks. r.t.	25,200	42,610	11.4
C2555	As above					4 hrs. 540°C, B.W.Q.* 6 hrs. 200°C	31,300	42,610	4.3
C2568	As above					4 hrs. 540°C, B.W.Q.* 6 hrs. 200°C	30,100	39,950	3.0

*Boiling water quench.

aluminum-copper alloy developed a tensile strength at least the equivalent of the aluminum-copper-beryllium-cobalt alloy with considerably higher elongation. The yield strength of the binary alloy is somewhat lower than that of the aluminum-copper-beryllium-cobalt alloy but it is obvious that in view of the relatively high elongation a precipitation treatment could be evolved which would give a higher yield strength and lower elongation. As indicated in the previously mentioned reference, even superior properties are obtainable in binary alloys of higher copper content. It must be concluded, therefore, that the addition of 0.2 per cent beryllium and 1.3 per cent cobalt to high purity aluminum-copper alloys does not appreciably enhance the mechanical properties of separately cast test bars.

The work we have done on the aluminum-copper-beryllium alloys confirms Dr. Harrington's conclusions regarding what he terms the superior thermal stability of the aluminum-copper-beryllium-cobalt alloys. In short, the presence of a few tenths of a per cent beryllium in aluminum-copper alloys has an appreciable effect on the precipitation characteristics of these materials. However, other commercial aluminum alloys, such as Y-alloy, 4.0 per cent copper, 2.0 per cent nickel, 1.5 per cent magnesium, have even superior thermal stability and their cost is considerably lower than that for an alloy containing 0.2 per cent beryllium and 1.3 per cent cobalt.

The important practical consideration has to do with the relative commercial casting characteristics of these high purity alloys as compared with such commercial materials as 195 and 356. Several attempts have been made in this country to exploit commercially high purity alloys of the aluminum-copper type with but little success. However, it is reported that considerable tonnages of castings are being made in England and France of a high purity aluminum-copper alloy to minimum tensile property specifications of the order of 40,000 pounds per square

inch with 5 per cent elongation. If the aluminum-copper-beryllium-cobalt alloy proves to be as commercially castable as 356 or even 195, it may find limited commercial application in spite of its relatively high cost. Unfortunately this question cannot be answered in the laboratory and the final answer will be forthcoming only after considerable plant experience.

Written Discussion: By M. G. Corson, metallurgical consultant, 603 West 111th Street, New York City.

I wonder whether this paper might not be called a report on an investigation. It does not cover a range of compositions, does not show why the specific ones were selected and contains no comparison data for alloys containing no beryllium. It amounts to a set of statements in regard to the characteristics of the selected alloys.

One of these statements mentions the necessity, or desirability, of having definite ratios between the weight percentages of copper or cobalt on one side and beryllium on the other. The formation of a compound CoBe is stated definitely; that of CuBe is indicated indirectly.

As far as CoBe is concerned the author states that its presence in the absence of copper does not yield anything of importance. This is as it ought to be because a compound of this type might form only a secondary independent phase and if present to the extent of 4 per cent act essentially as a ballast, slightly increasing the hardness of the alloy and sharply damaging its ductility.

However, no CoBe compound can form within an aluminum-base alloy. The reason is that in an alloy containing as much as 95 per cent aluminum against 3 per cent cobalt and 0.5 per cent beryllium, there are about 65 atoms of aluminum to 1 of beryllium and 1 of cobalt. Aluminum combines with cobalt with a great avidity to form the compound $CoAl_4$ so that each atom of beryllium would have to compete for an atom of cobalt against sixteen times the theoretical amount of aluminum needed. I think it would be a miracle if as much as 0.10 per cent beryllium would combine with cobalt; the remaining beryllium if not the whole of it must remain as a free phase.

The same considerations apply in the case of copper. If CuBe could form directly from the molten alloy on solidification and if it stayed so it would form a ballast of a very limited usefulness. In order to actually harden the matrix CuBe must form by precipitation from a solid solution. Consequently the latter must form first and this is possible only when both components individually can enter into the solid solution in aluminum during the solubilizing treatment. Such, for example, is the case of the Mg_2Si compound. Beryllium does not enter into solid solution in aluminum to any appreciable extent and copper cannot pull it in. Consequently the hardening of the aluminum matrix through a precipitation of CuBe is out of the question. The prerequisite of the individual solubility of both components is a hard and inescapable rule and Dr. Harrington presented no proofs to show that he was able to beat that rule.

Now, suppose for the sake of argument that he did so and hardened his alloy by precipitating CuBe from the crystals of the matrix. What has been gained? The author shows that his alloys had a tensile strength (sand cast) of 44,000 pounds per square inch and 5 per cent elongation. A plain, but well made, alloy containing 4 per cent copper is almost certain to show 45,000 pounds per

square inch and 6 per cent elongation while 42,000 with 4 per cent can be guaranteed. Obviously the 0.5 per cent beryllium used in connection with Dr. Harrington's technique produces no tangible improvement.

I emphasize "Dr. Harrington's technique", because beryllium taken in the amounts stated must produce a definite, though a very modest, improvement if properly used. The effect expected should amount to + 3000 pounds per square inch in tensile strength, + 2 per cent in elongation and + 1000 pounds per square inch in yield strength. I will not have the time to discuss the theory of this effect, but might mention in passing that it begins when the beryllium content goes beyond the eutectic composition and the beneficial range does not go beyond 0.1 or 0.2 per cent additional.

This slight effect can be gained at the cost of 0.5 per cent beryllium plus the cost of making the master alloy with a copper base. At the present price of beryllium, which, by the way, shows no tendency to drop, it amounts to over 20 cents per pound of molten metal and to far more per pound of finished castings.

The author states that his alloys possess a high resistance to oxidation. Can he give reasons why a hardened matrix containing no beryllium (or even containing it) and accompanied by secondary constituents (beryllium plus $CoAl_4$ as is the case, or CoBe as the author states) might be specifically resistant to oxidation? The fact is that a fairly pure (0.5 per cent impurities) but well-made alloy containing 4 per cent copper and no beryllium is just as resistant to oxidation. I mean only that perfectly dry oxidation as takes place during the solubilizing treatment in a muffle furnace.

The technique used by Dr. Harrington is unnecessarily complicated. Heating the charge of aluminum to about 1500 degrees Fahr. (815 degrees Cent.) means extra time, extra fuel and extra absorption of gas. Freezing the melt means a great loss of time (about 2 hours in the case of a 500-pound charge). The degasifying effect is most incomplete, unless the whole charge is allowed to freeze and the upper crust is kept broken all the time. Finally there is little need for the preparation of the 12.5 per cent beryllium copper-base master alloy. The proper technique in the case of a 500-pound heat would be:

1. Make a 10 per cent beryllium-aluminum alloy. It forms easily at about 1800 degrees Fahr. (980 degrees Cent.).

2. Put all the master alloy needed (5 per cent of the charge) in the bottom of the crucible. Cover it with the charge of aluminum. The latter need not be better than 99.6 per cent pure (if that pure). Melt the charge. Stir vigorously and bring the temperature to 1350 degrees Fahr. (730 degrees Cent.), and keep it there.

3. Put in all the copper required, preferably in the shape of coiled, clean, heavy wire (master alloy useless); it will dissolve rapidly.

4. Pour the molten alloy in a preheated ladle of at least 50 pounds capacity. Send through it a lively stream of boron trichloride for 1 minute. Let the temperature drop to 1250 degrees Fahr. (675 degrees Cent.) (or 1280 degrees Fahr. if you wish to have 3 per cent cobalt in the alloy). Pour the metal.

4A. Or, if one is too stingy to use about 0.5 cents worth of boron trichloride per pound of the charge, one may use a stream of nitrogen for 2 minutes. It would cost you about 0.1 cent per pound of metal.

But, of course one does not need to add the beryllium at all. An ordinary 4 per cent aluminum-copper alloy will do just as well and will cost much less.

In short, my conclusions are as follows:

1. The theoretical concepts leading to the development of the aluminum-copper-beryllium and aluminum-copper-cobalt-beryllium alloys are quite wrong.

2. The characteristics of the alloys as produced by Dr. Harrington show no improvement in comparison to substantially binary aluminum-copper alloys.

3. The possible (but not shown by the author) improvement due to beryllium is inconsequential and the use of this element is (in aluminum-copper alloys) economically unsound.

4. Dr. Harrington's technique is needlessly complicated.

Author's Reply

First of all, the author wishes to thank the five gentlemen for the interest they have shown in their discussion of the paper. Secondly, for the reader's interpretation, it should be realized that this paper was prepared under wartime restraints, about a year before this convention. Both Germany and Japan were still fighting us. The reported information was rather largely limited to the basic development work, most of the test data having been taken 4 to 5 years ago. For example, relative to melting procedure, copper-beryllium master alloys were necessarily employed because aluminum-beryllium stock has been made available only very recently.

Mr. Edens' discussion is particularly appreciated since it is the only available testimony (other than from several General Electric foundries) relative to the "castability" and practical foundry properties of these new alloys.

Dr. Sawyer states that the author's new alloy No. 1, consisting of 3.8 per cent copper, 0.2 per cent beryllium, balance aluminum, "is suggestive of a German casting alloy known as Hydronalium No. 511 consisting of 5 per cent magnesium, 1 per cent silicon, 0.2 per cent beryllium, 0.3 per cent manganese, 0.15 per cent titanium, balance largely aluminum, wherein magnesium replaces the copper of (the author's) alloy." Actually these two alloys are similar only in that both contain 0.2 per cent beryllium. Beyond that they may be as unlike as a plain low carbon steel and a stainless steel of the same carbon content. It seems probable that in this Hydronalium 511 a beryllium-silicide forms one of the precipitating phases. As Dr. Sawyer has noted for the hydronalium, the addition of a small amount of beryllium does benefit the melts of the author's alloys in inhibiting dross formation and conferring liquidity. Results from several hundred melts indicate that, with good foundry practice, only about 0.03 per cent beryllium is lost to oxidation so that it would appear that an addition of 0.05 per cent beryllium should suffice for this purpose alone.

Mr. Bonsack has inferred from the paper (and not unreasonably) that the author's new alloys require 99.85 per cent or higher purity aluminum and consequently "twice the cost of most common high strength aluminum alloys." Mr. Bonsack's primary interest is in secondary alloys. Relative to impurities, the only present comparison is on the basis of virgin ingot of commercial alloys. Comparison with the specifications for casting alloy 195 and the relatively low cost wrought alloy 2S (commercially pure aluminum) appear to be as follows:

195 Class	Per Cent	2S Class	Per Cent
Silicon	1.2 Max.	Aluminum	99.00 Min.
Iron	0.8 Max.	Iron + Silicon	1.00 Max.
Magnesium	0.03 Max.	Zinc	0.10 Max.
Titanium	0.2 Max.	Copper	0.20 Max.
Zinc	0.1 Max.	Manganese	0.05 Max.
Manganese	0.3 Max.		

(Alloy 195, of course, contains 4 to 5 per cent copper.) In the 195 specification, it appears that the silicon, titanium and manganese are added deliberately for certain alloy properties and are not required for the author's alloys. Relative to the author's alloy No. 3 there seems to be no reason why virgin ingot should not be made at comparable melting cost with impurities (elements other than copper, beryllium and cobalt) as follows: 0.8 per cent iron max., 0.1 per cent zinc max., 0.05 per cent manganese max., 0.1 per cent silicon max. It is to be noted that these are maximum limits and the value of the new alloys might well merit some increase in cost to lower the iron content still further although this is not yet indicated as necessary. Silicon, which competes with cobalt for beryllium, is to be avoided but 0.1 per cent max. silicon would involve only about 0.02 per cent beryllium which is negligible. It is, therefore, not unreasonable to hope that the new alloys might be increased in cost only by the added metal cost of the beryllium content: for alloy No. 3, about 3 to 4 cents per pound (based on 5 per cent copper-beryllium master alloy). It may be that the cost will be lowered somewhat by the use of aluminum-beryllium master alloy.

As to Mr. Bonsack's interesting data on his company's Ternalloys: In the table headed "Sand Cast" he has listed the "0.01 per cent proportional limit" and has quoted thereunder the properties of the author's alloys. This is in error for the author's proportional limit data were taken with an accuracy of 0.001 per cent. Consequently, for accurate comparison, the proportional limits for alloys 1, 2, 3 and 4 should be increased by at least 5000 pounds per square inch. It is unfortunate that the Ternalloys have an elongation, sand cast, of below 3 per cent which is usually indicative of low shock resistance. Under the heading of "Chill Cast", Mr. Bonsack has listed corresponding properties of his Ternalloys, apparently comparing them with one set of properties for forged alloy No. 3. Reference to page 322 of the paper will disclose another set of properties for forged alloy No. 3 (another heat treatment) showing that alloy No. 3 has strength properties at least equivalent to chill cast Ternalloys and with much higher elongation. The basis for comparing chill cast Ternalloys with the forged new alloy No. 3 is not clear. Although he does not give the analyses for his several Ternalloys, Mr. Bonsack states that they are of the aluminum-magnesium-zinc type. There is today quite a family of aluminum-magnesium-zinc alloys, stemming from the old (1922) Fuller-Basch patent. Basically, these alloys have lower corrosion resistance. A small amount of chromium may be added to improve the corrosion resistance and about 0.2 per cent titanium may be added to refine the grain.

Mr. Bonsack refers to the last line of Table VI wherein, for alloy No. 3, the 0.2 per cent yield strength is very close to the tensile strength and yet the alloy shows 3 per cent of elongation. These data are from automatically recorded stress-strain curves. There is no known natural law that says that an alloy may not elongate (in the plastic range) under nearly constant load. We have in

our files quite a number of such curves for iron-base, copper-base, aluminum-base and zinc-base alloys.

Mr. Kempf's discussion is much appreciated because it corroborates the author's report of the unusual high strength properties that can be achieved with the new alloys. However, comparing Mr. Kempf's data with those for the author's alloy No. 3 in Table I of the paper, it will be seen that appreciably higher "elastic properties" (proportional limit and 0.2 per cent yield strength) can be obtained. This difference in tensile properties seems to be due to the fact that the author used room temperature 10 per cent salt solution for a quench whereas Mr. Kempf prefers quenching in boiling water which gives a somewhat slower rate of cooling in the quench. The author's data for alloy No. 3 (Kempf's No. C2568) also show considerably higher "elastic properties" than are obtained for pure 3.8 per cent copper binary aluminum-copper alloy (Kempf's No. C2556), for example: a *proportional limit* of 25,000 pounds per square inch for alloy No. 3 versus the 0.2 per cent yield strength of 25,200 pounds per square inch for Kempf's pure binary aluminum-copper alloy (C2556).

Mr. Kempf's discussion evolves three chief points: (a) the author's new alloys develop properties no better than the properties known for pure binary aluminum-copper alloys of same copper content. As shown in the preceding paragraph, without further detailed discussion, the available data do not support this contention. For example, the author's alloy No. 3 will yield a 0.2 per cent yield strength that is 36 per cent higher than the yield strength quoted by Kempf for the pure binary aluminum-copper alloy (C2556). Incidentally, for engineering use of the alloys, it is the proportional limit and yield strength that are of major importance, *not* the tensile strength.

(b) Mr. Kempf's discussion indicates that the pure binary aluminum-copper alloys cannot be produced in commercial foundries with the high properties given by the alloys produced in the laboratory and that the same may be true of the author's new alloys. That this is true for the high purity binary aluminum-copper alloys seems well borne out, since, in the United States, the commercial alloy of this type (alloy 195) possesses markedly lower properties even with the additions of 1.2 per cent silicon and 0.2 per cent titanium to make it of practical value. That the author's new alloys may not be so limited in actual practice is indicated in preceding discussion, especially by Mr. Edens: "No special practices were used and I feel sure that the alloys can be readily handled by any foundry possessing reasonable metallurgical and foundry control".

(c) The author's new alloys will be prohibitive in cost. This, of course, remains to be proved and is a subject affected by many factors. The subject has been dealt with somewhat, earlier in this discussion. The author's alloy No. 3 appears at present to have certain slight foundry and property advantages over the other new compositions. It is not at all impossible that the added cost of this new composition will prove to be the cost of the 0.2 per cent beryllium which may be added either in the form of 5 per cent beryllium-95 per cent copper or as one of the recently available aluminum-beryllium master alloys. This cost differential may prove to be only 3 or 4 cents per pound. This alloy also appears to have 100 per cent scrap value since remelts, consisting of 100 per cent first-melt

scrap, produce practically identical properties. This value would further lower the cost of this alloy in production.

Before considering Mr. Corson's discussion, brief remarks about the "precipitating phases" seem pertinent. In the series of new alloys, a total of three different phases occur as follows:

Alloy	Composition—Per Cent	Phases
1	3.8 Copper-0.2 Beryllium	A and B (about equal)
2	3.5 Copper-0.5 Beryllium	B predominant (slight A)
3	3.8 Copper-0.2 Beryllium-1.3 Cobalt	A + C

From the microstructures, as anyone can verify, A is definitely $CuAl_2$; for several reasons plus logic, B appears to be of the CuBe type but may involve aluminum in a ternary "compound"; C, definitely different from A or B, appears to be of CoBe type, possibly involving aluminum in a ternary "compound". The second phase that occurs in pure binary aluminum-beryllium alloys of low beryllium content (0.3 to 0.5 per cent) is definitely absent from the structures of these new alloys. For polished, unetched surfaces, phases A and B have similar types of geometry but are different shades of gray and are distinct to the eye. Phase C forms mostly in feathery platelets. After etching with 5 per cent hydrofluoric acid, examination by polarized light affords color distinctions as follows: Phase A is red to scarlet; phase B is a rich purple shading to dark gray; and phase C is a lighter gray lilac. Shades may vary with etching conditions and matrix effects on etching. Aside from its effects on properties, phases and microstructure, beryllium exerts a real alloying effect in these new compositions.

Now as to Mr. Corson's discussion, that gentleman appears to be in the position of trying to refute facts with "theories". The author wishes to make the following successive points with respect to his discussion:

If he were to carefully read pages 312, 313 and 314 of the paper, he must conclude that obviously quite a number of compositions were studied. However, only the most pertinent (useful) compositions that wholly describe this new family of alloys were reported in detail.

Mr. Corson cites an atomic ratio of 16 to 1 against the probability of formation of CoBe rather than $CoAl_4$. However, this is but one factor (and a lesser one). The diffusion reaction involves a logarithmic function and the logarithm of the ratio of 16 and 1 is greatly smaller than 16. If Mr. Corson will carry his calculations further he will find that the heat of formation of CoBe need be only more than 2800 (small) calories per mol greater than that for $CoAl_4$ and all of the cobalt will combine with beryllium. This is a very small difference in heats of formation which range about 30 to 40,000 calories per mol. Other theoretical considerations favor CoBe over $CoAl_4$. The best data for the maximum solid solubility of beryllium in pure aluminum indicate that such solubility (particularly in atom per cent) is sufficient for the reaction to go to completion.

Mr. Corson believes at least part, if not all, of the beryllium must occur as a "free phase". Such a phase, as it actually occurs in pure binary aluminum-beryllium containing 0.3 to 0.5 per cent beryllium, is entirely absent from the microstructure of the new alloys.

On the basis of his arithmetic ratio probability theory against the formation of CoBe, Mr. Corson would have to contradict the facts that the phase

CrBe is precipitated in the alloy of 99.5 per cent copper-0.1 per cent beryllium-0.4 per cent chromium and the phase CoBe is precipitated in the alloy of 97 per cent copper-0.4 per cent beryllium-2.6 per cent cobalt in preference (in both cases) to the precipitating phase CuBe which forms in the binary copper-beryllium system. The analogies to the new aluminum alloy compositions are identical.

These discussion points also apply to Mr. Corson's theory against the formation of the phase CuBe in these new aluminum alloys. A good theory must fit all the facts.

The superiority of the author's new alloys in tensile properties to pure binary aluminum-copper alloys of equal copper content has been demonstrated in the preceding discussion.

Mr. Corson's statement that the addition of beryllium in the author's new alloys "produces no tangible improvement" seems to be likewise refuted by the facts.

His undisclosed theories might prove very interesting if one had the time and opportunity to give them due consideration.

Mr. Corson seems to believe that beryllium in the form of 5 per cent beryllium-95 per cent copper (used for alloys No. 1 and 3) is more costly than making up beryllium-aluminum master alloy from metallic beryllium as he recommends. This certainly has not been the case. Throughout his discussion he has concentrated on the author's alloy containing 0.5 per cent beryllium and 3.5 per cent copper which would require the higher cost 12.5 per cent beryllium-copper master alloy. From a practical and application viewpoint, alloy No. 3 appears to be of greater value and this choice, fortunately, alters the economics considerably. It should also be obvious to any reader that, whenever the aluminum-beryllium master alloys prove more economical, they will naturally be used.

His remarks on oxidation resistance are not backed by any data and some of them are not clearly understood by the author. It is to be noted that the brief comments by the author in the paper, concerning this property, are supported by the other discussers.

Mr. Corson's alternative single-melting method is perfectly sound, of course. The author probably made a mistake in assuming that this would be obvious for foundries whose melters are trained to be that careful. There are many foundries wherein these new alloys would be better introduced (at the beginning) with the insurance guaranteed by the double-melting method.

His melting technique seems to imply that the melter would also make either the copper-beryllium or aluminum-beryllium master alloys. This is not the case, of course. There are approved sources of supply for these master alloys.

Mr. Corson recommends passing boron trichloride through the melt. So far as the author is concerned, this should definitely *not* be done. The boron acts as a deoxidizer and the chlorine is very apt to remove the beryllium from the alloy just as it does magnesium. Aside from this imperative danger in the use of boron trichloride, there is absolutely no reason for using it or nitrogen gas for these new alloys.

It is to be hoped that anyone wishing to try these new alloys will consult the

author relative to methods of alloying, etc. Since these new alloys are subject to patents (as noted in the paper, page 312, footnote) anyone interested in these alloys is guaranteed full co-operation from the author.

As to the four conclusions to Mr. Corson's discussion, the author concludes as follows:

(a) He shows no evidence of knowledge of the "theoretical concepts leading to the development of the aluminum-copper-beryllium and aluminum-copper-beryllium-cobalt alloys". The author has had no opportunity to inform him as to what these concepts might have been.

(b) The author's data prove conclusively that the new alloys have properties superior to "substantially binary aluminum-copper alloys" and Mr. Kempf's data attest to the same.

(c) If marked improvement in foundry properties and a 66 per cent increase in elastic properties (etc.) can be termed "inconsequential", then, and then only, can Mr. Corson be correct in stating that "the improvement due to beryllium is inconsequential". The engineering application and the user of the alloy will determine whether the increase in metal cost is justified.

(d) Whatever Mr. Corson may mean in several different references to "Dr. Harrington's technique", it may be safely assumed that the author will back up that "technique"!

THE SUPPRESSION OF PEARLITE IN MANGANESE-MOLYBDENUM STEELS

By Charles R. Austin and J. R. Doig

Abstract

This paper determines the approximate effects of carbon, manganese, and molybdenum on the cooling rate required to prevent formation of pearlite in hypoeutectoid manganese-molybdenum steels, taking as a standard of reference the formula of Grange and Kiefer.

S-curves of nine manganese-molybdenum steels are presented.

After the assumption is made that the effect of these elements on the pearlite reaction at 1200 degrees Fahr. (650 degrees Cent.) approximates their effects on the pearlite nose of the S-curve, the effects of carbon and molybdenum in molybdenum steels and of manganese in carbon steels are determined from the data of previous investigators by a new method. It is then found that the effects are unchanged when the elements are present in manganese-molybdenum steels. This observation allows the derivation of an approximate formula for pearlite-suppressing cooling rates as a function of manganese and molybdenum contents. A method is developed for comparing the limiting cooling velocity as calculated from Grange and Kiefer's formula with total alloy content.

This paper may also be considered as a study of the effects of manganese and molybdenum on the isothermal transformation time for pearlite at 1200 degrees Fahr. (650 degrees Cent.).

Introduction

KNOWLEDGE of the dependence of the rate of pearlite formation upon alloy content is important for annealing and heat treating. In carbon steels the austenite which does not transform to ferrite and pearlite on cooling through the pearlite forming tem-

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perature range will transform to martensite at lower temperatures. However, in steels alloyed sufficiently with carbide forming elements to produce two temperature regions of maximum transformation rate—one around 1200 degrees Fahr. (650 degrees Cent.) where pearlite is formed, and the other around 900 degrees Fahr. (480 degrees Cent.) where high temperature bainite is formed—this austenite will transform to either bainite or martensite or both. If the pearlite reaction is retarded much more than the bainite reaction, then pearlite may be suppressed easily, and a wide variety of microstructures can be produced by variation of the cooling rate. The pearlitic reaction can be retarded greatly in manganese-molybdenum steels without large alloy additions as is shown in isothermal transformation diagrams to be presented in this paper. Thus the rate of the pearlite reaction is especially interesting in this type of steel. It is the purpose of this paper to evaluate the effects of carbon, manganese, and molybdenum on the limiting cooling velocity required for suppression of pearlite in manganese-molybdenum steels. Little attention will be paid to the ferrite reaction.

A simple method of calculating from S-curve data the cooling rate necessary to prevent formation of a given structure which has a nose or temperature of maximum transformation rate has been presented by Grange and Kiefer (1)[1]. Their formula for the rate is:

$$R = \frac{Ae_n - T_n}{1.5 \times t_n} \qquad \text{Equation 1}$$

where Ae_n is the highest temperature at which it is possible to form the given structure, T_n is the nose temperature at which minimum transformation time occurs for the structure, and t_n is the time required for the structure to begin to form at the temperature T_n of the nose. If a definite pearlite nose exists in the S-curve, this formula may be made to apply to the suppression of pearlite by substituting for Ae_n the Ae_1 temperature, for T_n the temperature T_{pn} of the pearlite nose, and for t_n the time t_{pn} for the beginning of isothermal formation of pearlite at the nose. The formula, as thus applied to the suppression of pearlite, will be used in this paper as an approximation or standard of reference for determining from isothermal transformation studies the effect of carbon, manganese, and molybdenum on the rate of cooling required for suppression of pearlite.

[1]The figures appearing in parentheses pertain to the references appended to this paper.

While the temperatures Ae_1 and T_{pn} vary with chemical composition, their difference which is the entire numerator of equation 1 varies much less because the alloying elements which raise the Ae_1 temperature also raise the temperature of the nose. The time t_{pn} for the beginning of transformation at the nose is by far the most important part of the expression, for it varies in published S-curves from less than a second for plain carbon steels to over 2000 seconds in S.A.E. 4340 steel.

The average temperature of the pearlite nose in S.A.E. and NE steels whose published S-curves contain bays is close to 1200 degrees Fahr. (650 degrees Cent.). Only with very high alloy content or with alloy content too low to produce a bay in the S-curve is the temperature of the pearlite nose moved much more than 50 degrees Fahr. from the 1200 degree Fahr. (650 degree Cent.) level. Therefore, the effect of alloy elements on the time for beginning of pearlite formation at 1200 degrees Fahr. (650 degrees Cent.) will approximate the effect on the time position of the nose. This temperature level is advantageous for study also for the reasons that it is above the levels at which either bainite or the X-constituent appear, and that it is a level at which direct measurements of the start and end of reactions are always made in constructing an isothermal transformation diagram.

In the following section of this paper, the S-curves of several manganese-molybdenum steels will be presented. In preparation for an analysis of the effect of carbon, manganese, and molybdenum on the time for beginning of pearlite formation at 1200 degrees Fahr. (650 degrees Cent.) in these steels, a quantitative study of the effects of each of these elements separately will be made from the data of Blanchard, Parke, and Herzig (2), (3) and of Davenport (4). With the aid of this information, the effects of manganese and of molybdenum in the manganese-molybdenum steels of the present authors will be determined and compared with the effects of the elements singly. The limiting cooling rate for pearlite suppression (based on an approximation of Grange and Kiefer's formula) will then be stated in a fully numerical expression involving manganese and molybdenum contents. A method will later be developed for comparing the limiting cooling rate calculated by the direct use of Grange and Kiefer's formula from S-curve data with the total alloy contents of steels. These results will then be compared with the numerical equation.

Experimental Procedure

Materials: Seven manganese-molybdenum steels were obtained from the Copperweld Steel Company, and two were made in the Metallurgy Laboratories of The Pennsylvania State College. All the steels were aluminum-killed. Compositions and grain sizes are listed in Table I.

Table I
Compositions of Steels Investigated

Steel Number	Composition, Per Cent						ASTM Grain Size
	C	Mn	Si	Mo	Ni	Cr	
38	0.27	0.86	0.25	0.71	0.05	0.03	7–8
42	0.29	1.64	0.24	0.45	0.02	0.05	7–8
13	0.39	1.10	0.71	0.46	...	...	9
15	0.43	0.76	0.27	0.67	...	...	10
28	0.39	1.58	0.27	0.44	0.05	0.16	8–9
27	0.50	1.50	0.17	0.45	0.10	0.20	7–8
30	0.52	1.18	0.30	0.30	0.16	0.13	8–9
29	0.80	0.90	0.30	0.44	0.16	0.09	8–9
31	0.72	0.32	0.19	0.45	0.09	0.09	7–8

Steels No. 13 and No. 15 were made in The Pennsylvania State College Metallurgy Laboratories. All others were furnished by the Copperweld Steel Company.

Heat Treatment of Specimens for S-curve Study: Specimens 0.045 to 0.065 inch thick were cut transversely from bars 5/8 inch square, which had been normalized from 1740 degrees Fahr. (950 degrees Cent.). After austenitizing at 1560 degrees Fahr. (850 degrees Cent.) for 25 minutes, the specimens were quenched into a salt bath at the desired temperature and allowed to transform for given lengths of time before being quenched in water to stop the transformation. The austempering bath was held to within 6 degrees Fahr. of the desired temperature.

Austempering times were 5, 10, 100, 1000, 10,000 100,000 seconds at temperatures of 1300, 1200, 1100, 1000, 900, and 800 degrees Fahr. (705, 650, 595, 540, 480, and 425 degrees Cent.). In some instances temperatures of 1050 and 1175 degrees Fahr. (565 and 635 degrees Cent.) were also used. The specimens transformed as indicated above were examined metallographically.

The course of transformation was determined from a consideration of the per cent transformation of the various specimens which allowed the construction of approximate reaction rate curves. From these curves (not reproduced) the times for beginning and end of transformation were estimated. It is believed that in this way the times for the start and finish of the pearlite reaction were fixed

to ±1/10 of a log time interval. For the 1200 degree Fahr. (650 degree Cent.) temperature level additional austempering times were used to fix more closely the beginning and ending times for the pearlite reaction.

Since the metallographic data did not appear to be adequate to completely describe the course of the bainite reaction, supplementary dilatometric data were obtained. Good agreement for the start of reaction was obtained in the two methods, but the dilatometrically determined end of bainite formation invariably preceded that determined metallographically. This difference is accounted for mainly by the facts that the precipitate structure is acicular and that a very small amount of austenite is retained up to very long transformation times in some instances. In such cases it is difficult to construct from metallographic data the reaction rate curves, which are used to determine the end of transformation. Dilatometer studies were not made on steels 13, 15, 27, and 38 except at 800 degrees Fahr. (425 degrees Cent.), but the metallographically determined end of bainite reaction was taken at the practical end of the transformation.

Dilatometer specimens were austenitized 3 minutes at 2010 degrees Fahr. (1100 degrees Cent.). (One steel was studied for austenitizing times of 1, 2, 4, 8, and 16 minutes at 2010 degrees Fahr. (1100 degrees Cent.) with subsequent determination of the limits of the bainite reaction.) The various austenitizing times had no appreciable effect on the limits of the bainite reaction. The specimens for dilatometric study were heated in a vertical tube furnace under an atmosphere of purified nitrogen. Transfer time for the dilatometer specimens (a modified Davenport-Bain (5) type dilatometer was used) was less than 5 seconds. All specimens were at a bright-red heat on immersion in the lead quenching bath.

The austenite grain size was determined using Vilella's grain size etchant on quenched and tempered specimens which had been austenitized at 1560 degrees Fahr. (850 degrees Cent.) for 25 minutes.

S-curves: Isothermal transformation diagrams of the nine steels studied are shown in Figs. 1 through 9.

Some qualitative observations may be made from even a cursory examination of the curves. In steels 42, 28, and 27 where carbon is the principal variable, the ferrite and bainite reactions are markedly retarded by increase of carbon content, but the pearlite reaction remains relatively stable. In the steels containing 0.45 per cent molybdenum, namely steels 31, 29, 13, 27, 28, and 42, the pearl-

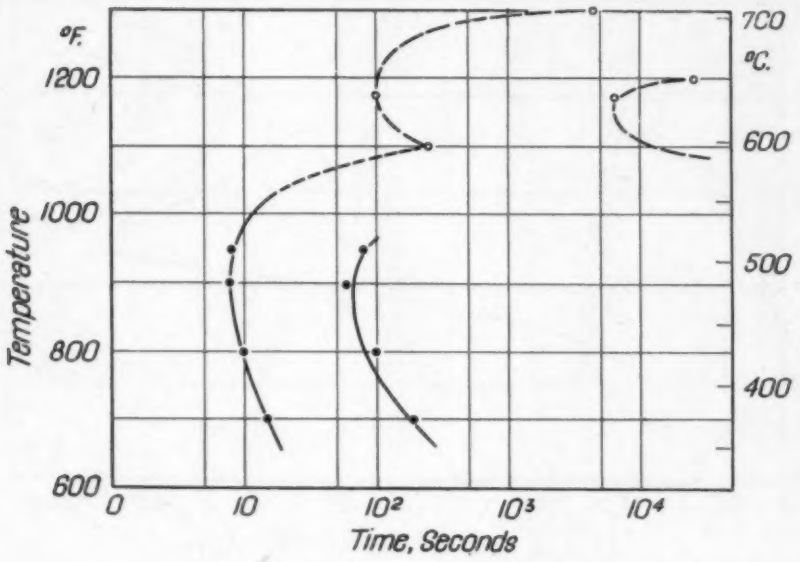

Fig. 1—S-curve for Steel 42. Carbon 0.29 Per Cent, Manganese 1.64 Per Cent, Molybdenum 0.45 Per Cent.

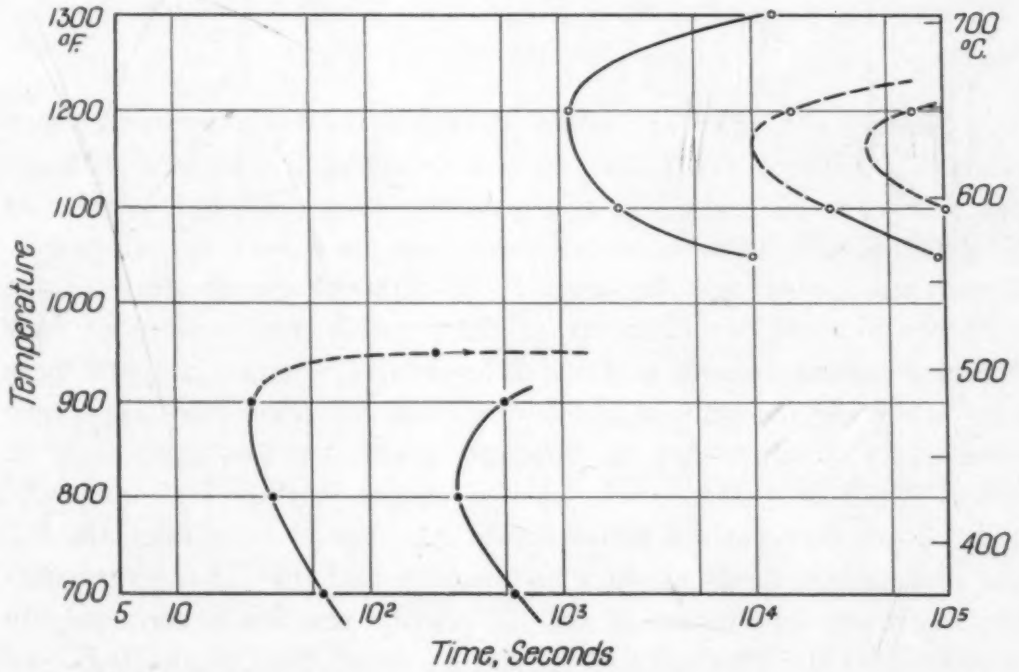

Fig. 2—S-curve for Steel 28. Carbon 0.39 Per Cent, Manganese 1.58 Per Cent, Molybdenum 0.44 Per Cent.

ite nose is progressively pushed to the right or retarded relative to the bainite nose position, illustrating the much greater retarding effect that manganese has on pearlite than it has on bainite.

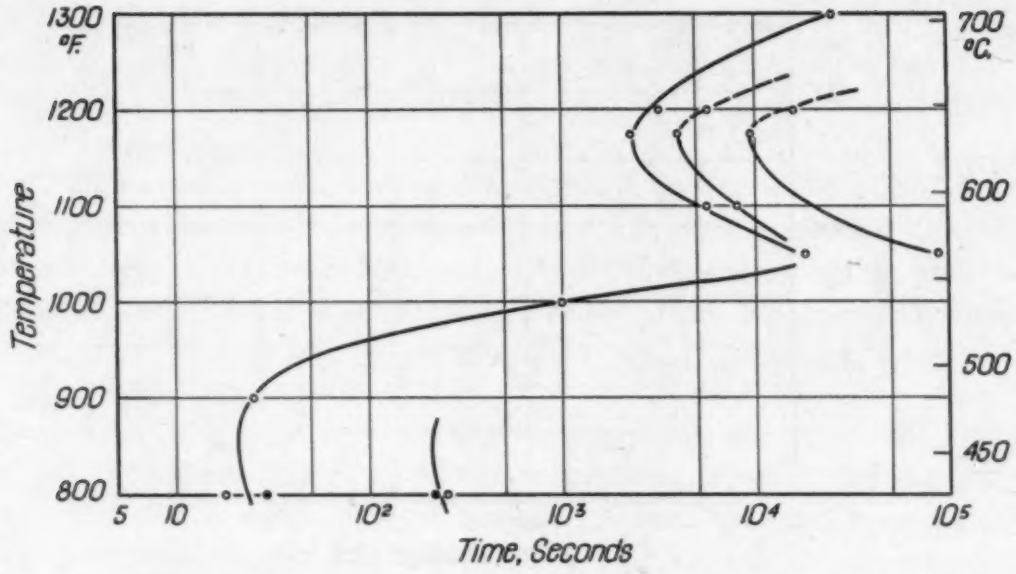

Fig. 3—S-curve for Steel 27. Carbon 0.50 Per Cent, Manganese 1.50 Per Cent, Molybdenum 0.45 Per Cent.

Quantitative Effects of Carbon and Alloying Elements on the Time for Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.)

Sources of Data: The effect of carbon on the S-curve has been studied by Davenport (4) for three different carbon ranges. Blanchard, Parke, and Herzig (2), (3) moreover have published a series of S-curves of molybdenum steels from which the effects of both molybdenum and carbon may be seen. In this comprehensive study of the influence of these two elements, all the possible combinations of four different carbon contents and five different molybdenum contents were used while the manganese and silicon concentrations were kept very close to fixed values and the austenite grain size was maintained at No. 8 A.S.T.M. Of the four carbon ranges in their series, the 0.1 and 1.2 per cent carbon series are of no interest here since the 0.1 per cent carbon steels produce no pearlite, and the 1.2 per cent carbon steels are hypereutectoid and the pearlite reaction is consequently nucleated by the proeutectoid cementite, while the 0.4 and 0.77 per cent carbon ranges are of interest in this study of hypoeutectoid steels. The effect of molybdenum on carbon steels has also been studied by Davenport (4), and this same author has studied the effects of manganese, chromium, nickel and other alloying elements. Russell and McGuire (6) have studied the effect of manganese, principally at high concentrations.

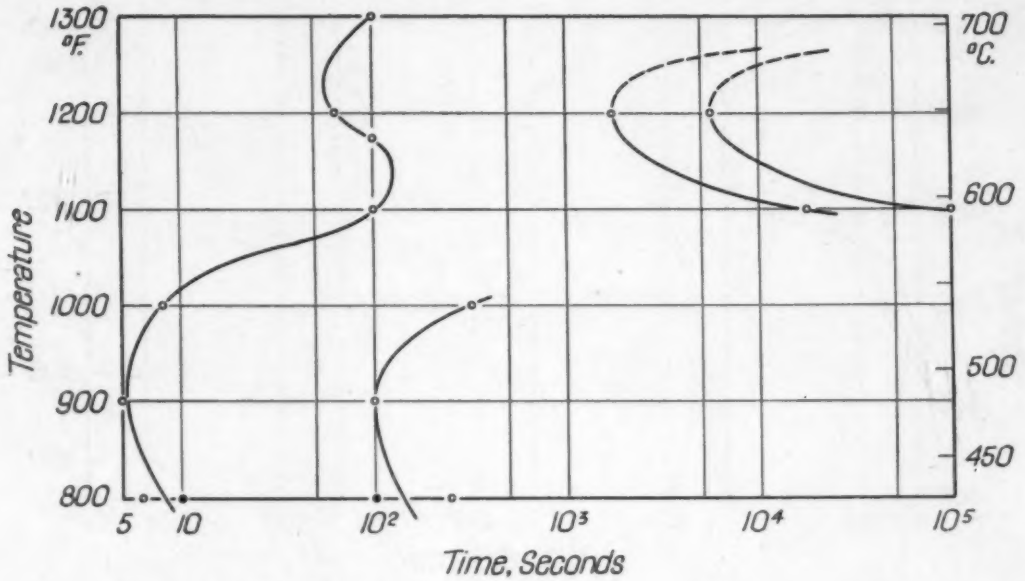

Fig. 4—S-curve for Steel 13. Carbon 0.39 Per Cent, Manganese 1.10 Per Cent, Silicon 0.71 Per Cent, Molybdenum 0.46 Per Cent.

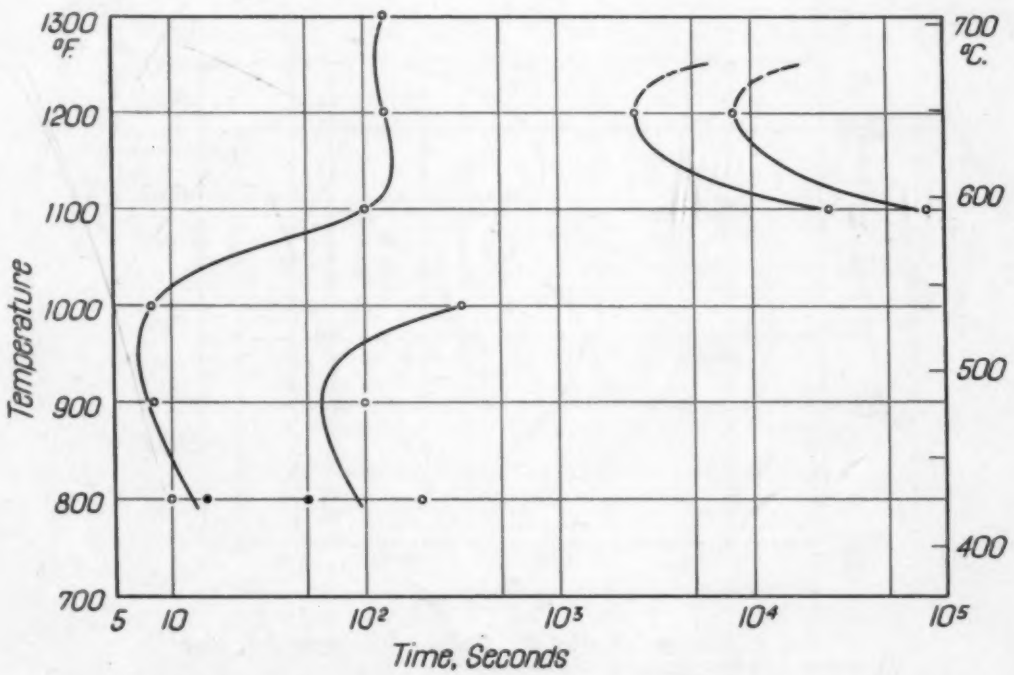

Fig. 5—S-curve for Steel 15. Carbon 0.43 Per Cent, Manganese 0.76 Per Cent, Molybdenum, 0.67 Per Cent.

None of these authors has published a quantitative analysis of the effect of any element. From their data, therefore, which have heretofore been examined only qualitatively, an analysis will be made graphically here to show the effect of each element singly in preparation for a study of its effect in manganese-molybdenum steels.

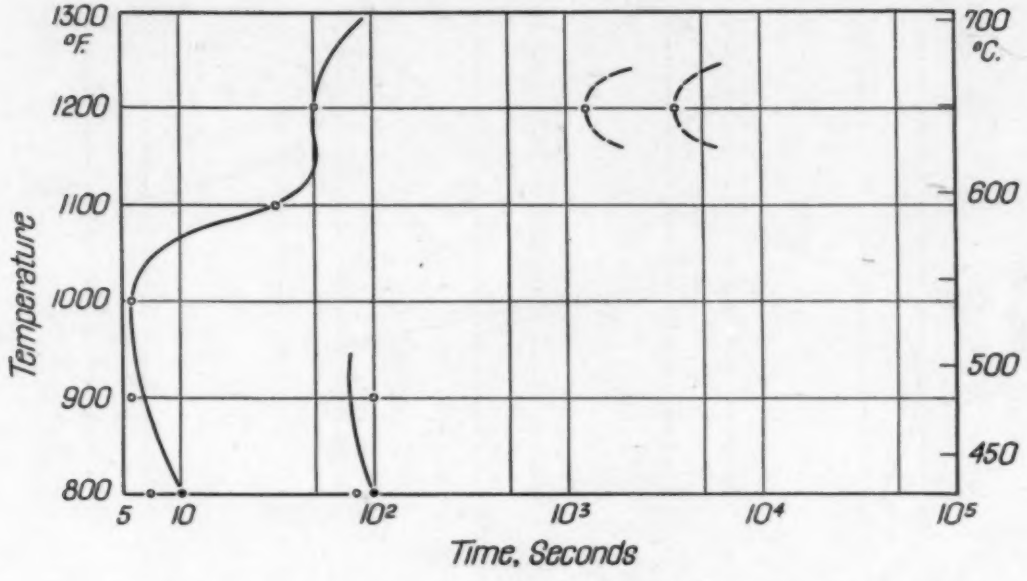

Fig. 6—S-curve for Steel 38. Carbon 0.27 Per Cent, Manganese 0.86 Per Cent, Molybdenum 0.71 Per Cent.

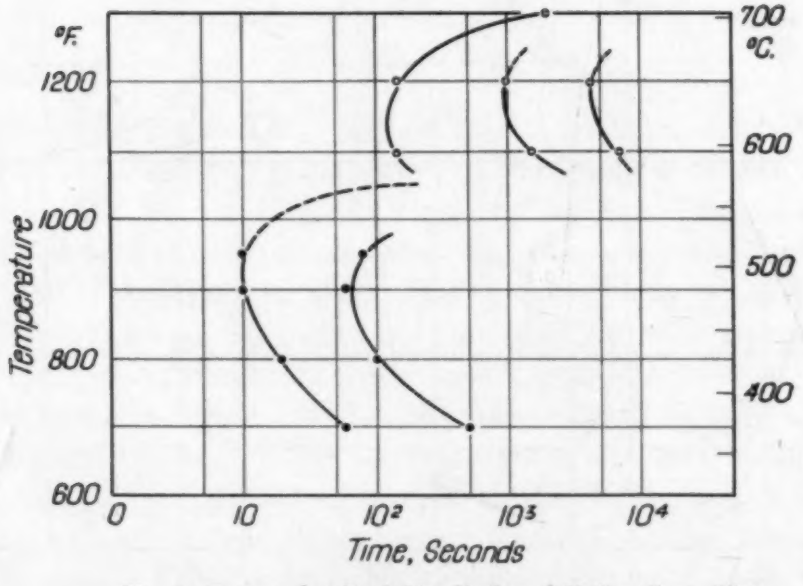

Fig. 7—S-curve for Steel 30. Carbon 0.52 Per Cent, Manganese 1.18 Per Cent, Molybdenum 0.30 Per Cent.

Effect of Carbon on Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.): From the papers of Blanchard, Parke, and Herzig (2), (3), S-curves of two series of hypoeutectoid steels of grain size No. 8 A.S.T.M. and with varying molybdenum contents are available for comparison with each other. In one series, carbon is maintained at 0.39 to 0.45 per cent while in the other, carbon con-

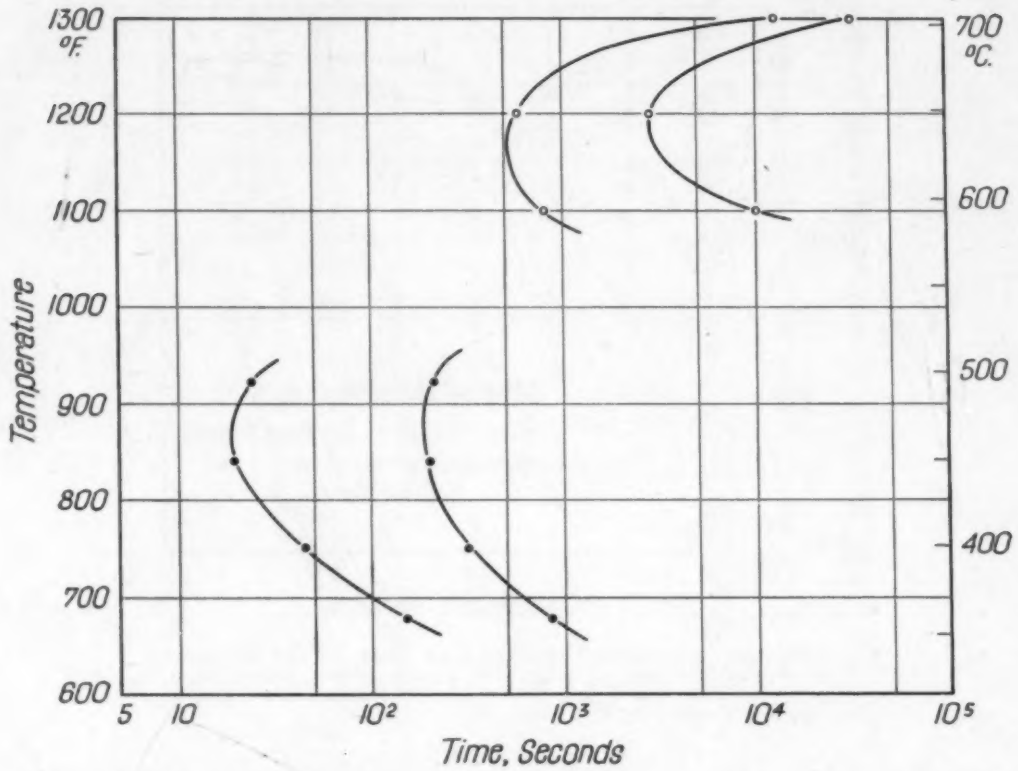

Fig. 8—S-curve for Steel 29. Carbon 0.80 Per Cent, Manganese 0.90 Per Cent, Molybdenum 0.44 Per Cent.

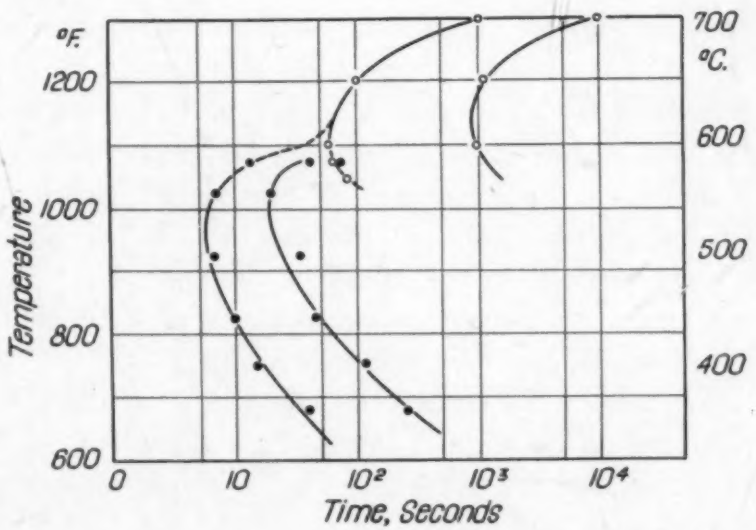

Fig. 9—S-curve for Steel 31. Carbon 0.72 Per Cent, Manganese 0.32 Per Cent, Molybdenum 0.45 Per Cent.

tents are held between 0.77 and 0.79 per cent. While the latter range is nominally called "eutectoid" by the authors, it is recognized to be hypoeutectoid for the purposes of the present study because of the presence of free ferrite in photomicrographs of the steels fully trans-

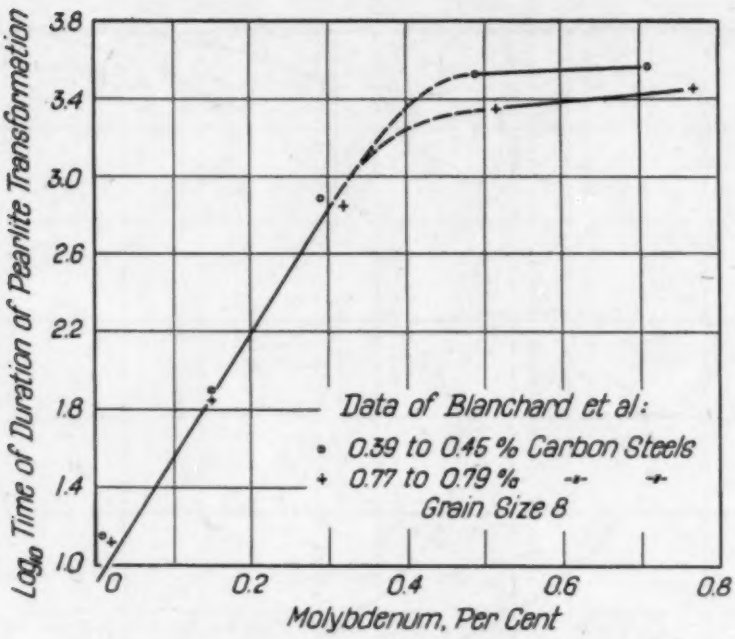

Fig. 10—Isothermal Transformation Data for the Duration Time of Pearlite Formation at 1200 Degrees Fahr. according to Blanchard, Parke, and Herzig (2), (3) after Correction to a Constant Manganese Content of 0.70 Per Cent.

formed at 1200 degrees Fahr. For an examination of the effect of carbon in these two ranges, the logarithm of time elapsed between beginning and ending of pearlite formation is plotted against molybdenum content in Fig. 10. Here the position of each point on the log time ordinate is corrected to the position that should be expected if the steel contained 0.70 per cent manganese. This correction was made on the basis of the observation from Davenport's data (4) that manganese increases the logarithm of the transformation time in seconds at 1200 degrees Fahr. at the rate of 1.6 per one per cent manganese (see section on the effect of manganese). The corrections are not large, for the variations in manganese concentration in these steels are small; but they are appreciable.

The logarithms of the transformation times for the 0.4 per cent carbon series tend to be slightly larger than those for the 0.77 per cent carbon series, but the difference is not important except at 0.5 per cent molybdenum and above.

A similar presentation of the data for the logarithm of the time at which pearlite begins to form according to the same S-curves is plotted in Fig. 11. In this series of points also there is a slight tendency toward longer transformation times in the 0.4 per cent carbon steels. However, the tendency is so small at 0.3 per cent molybdenum and below that it amounts to an excess of only 25 per cent

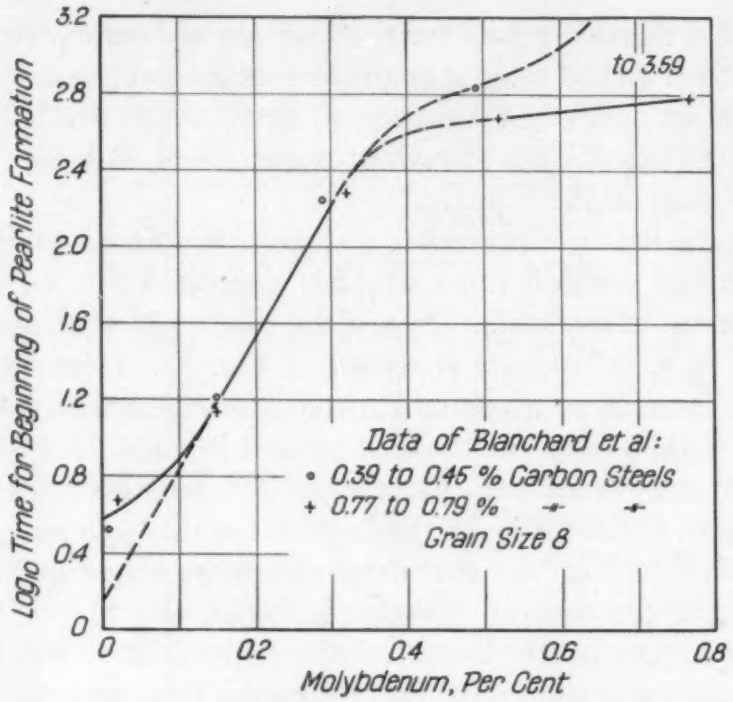

Fig. 11—Isothermal Transformation Data of Blanchard, Parke, and Herzig (2), (3) for the Time for the Beginning of Pearlite Formation at 1200 Degrees Fahr. after Correction to a Constant Manganese Content of 0.70 Per Cent.

more time on the average. For practical purposes it may be considered that the effect of carbon on the pearlite reaction rate at 1200 degrees Fahr. (650 degrees Cent.) in these steels is insignificant. (The slight difference in reaction time is equivalent to a molybdenum addition of about 0.015 per cent.) This observation should be considered to hold true only for a hypoeutectoid steel.

Effect of Molybdenum on Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.): In Fig. 10 a straight line can be drawn fairly close to the sets of points for the three lowest molybdenum concentrations. The slope of this line is about 6.5 per one per cent molybdenum. Though at higher molybdenum contents the curve appears to approach an asymptote, at concentrations up to at least 0.3 per cent there appears to be a practically linear relation between the logarithm of the transformation time and the molybdenum concentration.

In Fig. 11 the best curve that can be drawn through the first three sets of points is not linear, but if a straight line is drawn between the set for 0.15 and 0.30 per cent molybdenum, its slope is 7.1, which value is close to the value of 6.5 calculated for the curve for the duration of the transformation in Fig. 10. It is quite

probable that the two points for the lowest-molybdenum steels would also have been on the same straight line as the next two sets if their transformation times had not been so short and if the 1200 degree Fahr. (650 degree Cent.) level in these steels had not been well above the nose of the S-curve.[2]

From the data of Davenport (4) also the effect of molybdenum on the pearlite reaction times at 1200 degrees Fahr. (650 degrees Cent.) can be determined. These data, plotted in a manner similar to that in Figs. 10 and 11, is shown in Fig. 12. Here the reaction times are corrected to a constant manganese content of 0.4 per cent by the same ratio used for the data plotted in Figs. 10 and 11. If a line is drawn between the first two points for the transformation duration time, the slope is 6.7, while that for the beginning of transformation is 7.3. The fact that these values are very similar to those obtained from the data of Blanchard, Parke, and Herzig (2), (3) shows that a very nearly linear relationship exists between the logarithm of the nucleation and transformation time and the molybdenum concentration up to molybdenum contents of 0.3 per cent; otherwise the different molybdenum concentrations used by the two sets of authors could not have produced the similar values for the slopes.

The value of the slope of these curves is indicative of the rate of retarding of the pearlite transformation produced by the addition of the alloy element. It will facilitate dealing with this property if it is given a name. Let the slope of the curve of the logarithm (to the base ten) of the transformation or nucleation time against percentage added element be tentatively called the "retarding parameter". The foregoing paragraphs have investigated the retarding parameter for molybdenum on pearlite at 1200 degrees Fahr. (650 degrees Cent.). (The retarding parameter at other temperatures should be expected to be different, as should the retarding parameter on other structures such as bainite or ferrite.) It has been shown to be constant for molybdenum on pearlite up to about 0.30 per cent addition and to decrease with continued increase in molybdenum. The fact that it is constant or that the log time-percentage concentration curve is linear up to 0.3 per cent means that each small increment of molybdenum concentration produces the same *percentage* change in transformation time as any other equal increment until

[2]A specimen quenched to a temperature where the transformation time decreases rapidly with decrease in temperature will transform more slowly than it would if quenching were instantaneous because it spends the first part of the time in the bath at levels where its transformation rate is slow. This phenomenon is important only for measured times of a few seconds or less.

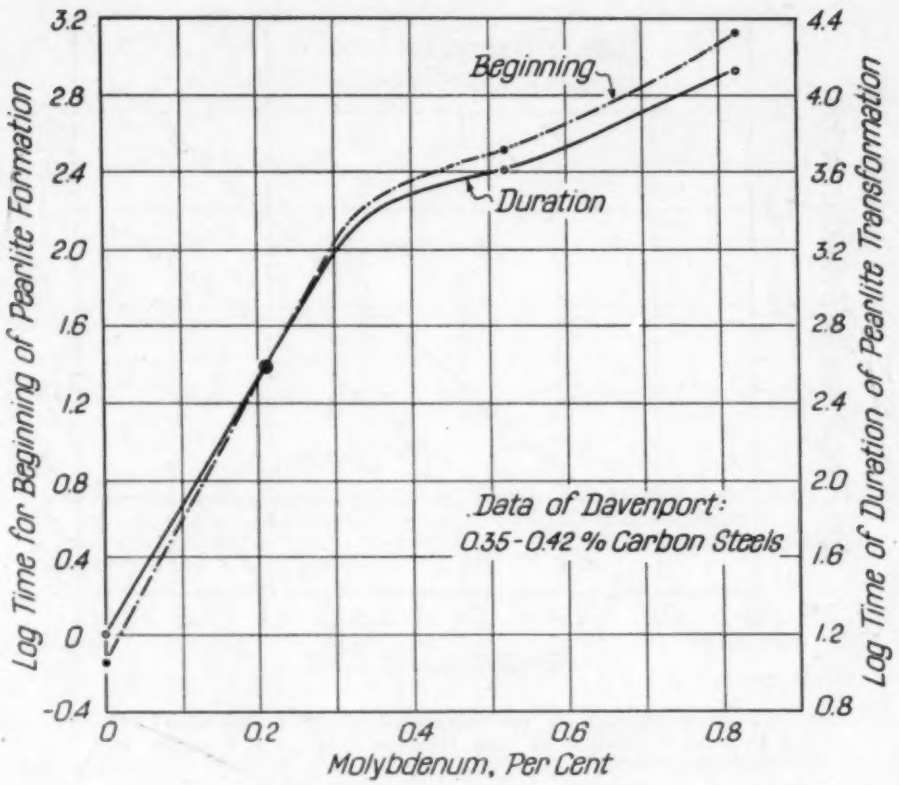

Fig. 12—Isothermal Transformation Data of Davenport (4) for Pearlite Formation at 1200 Degrees Fahr. as a Function of Molybdenum Content. Transformation Times are Corrected to a Constant Manganese Content of 0.40 Per Cent.

the total concentration is 0.3 per cent. In general, the retarding parameter should not be expected to be linear. In mathematical representation,

$$\frac{d \log_{10} t}{dc} = a \qquad \text{Equation 2}$$

where t is the transformation time, a is the retarding parameter, and c is the concentration in weight per cent.

The total change produced in the logarithm of the transformation or nucleation time by addition of C per cent of an alloy element is

$$\Delta \log_{10} t = \int_0^C a \, dc \qquad \text{Equation 3}$$

If the retarding parameter a is constant (log time vs. concentration curve linear) over any range of concentration of interest,

$$\Delta \log_{10} t = a \, (c_2 - c_1) \qquad \text{Equation 4}$$

and in such a case, calculations are simplified.

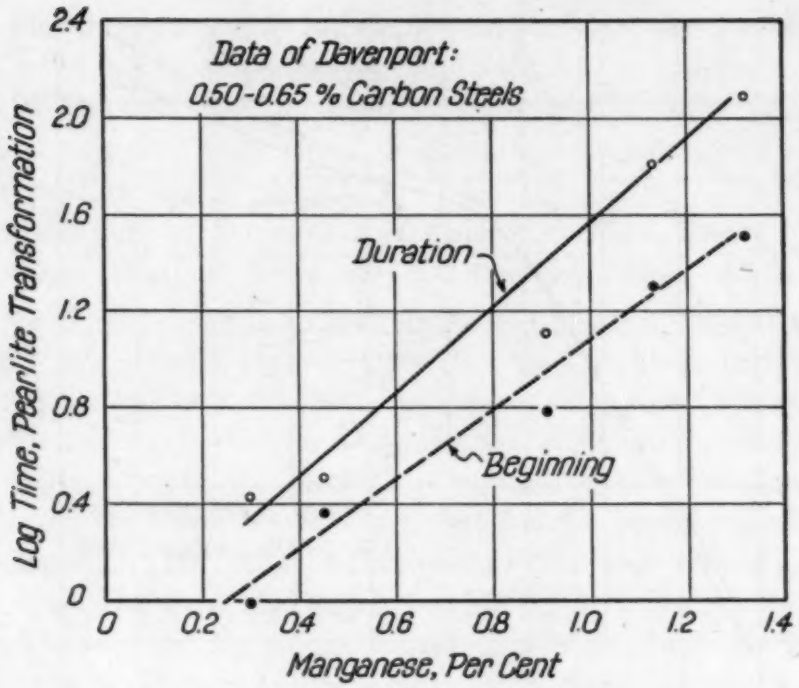

Fig. 13—Isothermal Transformation Data of Davenport (4) for Pearlite Formation at 1200 Degrees Fahr. as a Function of Manganese Content.

Effect of Manganese on Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.): The effect of manganese at 1200 degrees Fahr. (650 degrees Cent.) on the time for the beginning of pearlite formation and on the time for the duration of the transformation in plain carbon hypoeutectoid steels as determined from the data of Davenport (4) is shown in Fig. 13. Here there is no correction for the effect of other elements since manganese is presumably the only alloying element present. The retarding parameter for manganese calculated from the curve for the time for the beginning of transformation is 1.44, assuming that the curve is linear over the range of manganese concentration shown, while that calculated from the curve for the transformation time is 1.74. For reasons stated in footnote 2, the value 1.44 for the beginning of transformation may be lower than the true value. The curve appears to be not truly linear but slightly concave upward; nevertheless, the 1.44 and 1.74 values represent fairly true average values of the retarding parameter over the range from 0.3 to 1.4 per cent. Unfortunately, the grain sizes of these steels are not known.

Similar calculations of the retarding parameter taken from the data of Russell and McGuire (6) (not shown) yield 1.39 and 1.57

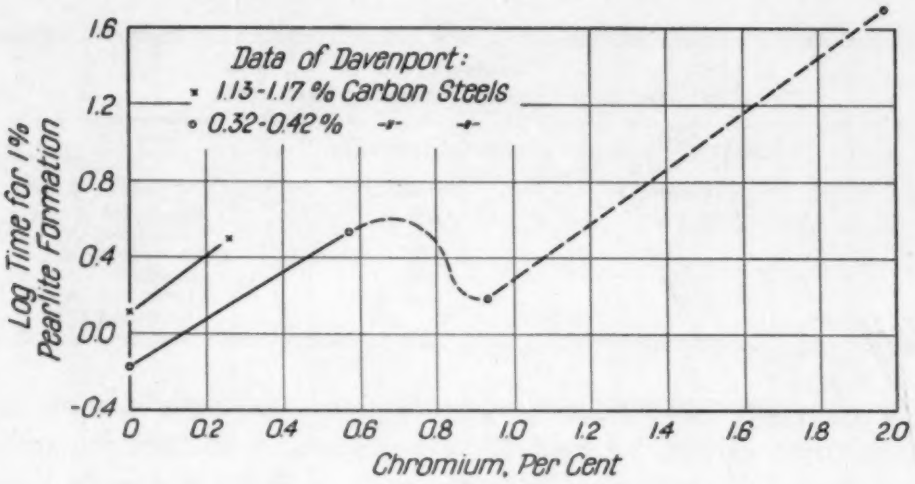

Fig. 14—Isothermal Transformation Data of Davenport (4) for the Time for the Beginning of Pearlite Formation at 1200 Degrees Fahr. as a Function of Chromium Content. Transformation Times are Corrected to Constant Manganese Contents in Each Curve.

respectively. However, these results are from data on manganese concentrations of 1.55 and 2.10 per cent. The average value of 1.6 from the data of Davenport therefore probably represents the most reliable value of the retarding parameter for manganese to be found from the existing data.

Determinations of the retarding parameter for manganese at other temperatures calculated from the data of Davenport (4) would reveal that manganese is more effective above 1200 degrees Fahr. and less and less effective at temperatures below.

Effects of Chromium and of Nickel on Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.): Davenport's data (4) on the effect of chromium on the time for beginning of pearlite formation at 1200 degrees Fahr (650 degrees Cent.) corrected to constant manganese concentrations is shown in Fig. 14. The effect of this element in first increasing and then decreasing the time for pearlite formation at 1200 degrees Fahr. (650 degrees Cent.) can be explained by the fact that chromium raises the Ae_1 point of a steel and also the pearlite nose. Though the nose is continuously pushed to the right by successive additions of chromium, it is also raised toward higher temperatures. Thus, if the 1200 degree Fahr. (650 degree Cent.) level in the S-curve of a steel of chromium concentration *A* is on the nearly horizontal portion of the upper side of the pearlite nose, an increase of chromium to concentration *B* may move the pearlite nose upward enough to actually produce a decrease in transformation or nucleation time. Thus the method of determin-

Table II
Tabulation of Data on Retarding Parameter

Element	Retarding Parameter at 1200 Degrees Fahr. (Per One Per Cent) Beginning	Duration	Source of Data
Mo	7.1	6.5	Blanchard et al. (2), (3)
	7.3	6.7	Davenport (4)
Mn	1.44	1.74	Davenport (4)
	1.39	1.57	Russell & McGuire (6)
Cr	1.4	...	Davenport (4)
Ni	0.55		Davenport (4)

ing the retarding parameter previously used for molybdenum and for manganese cannot be used for chromium. A method of studying the change in the transformation time at equal distances below the Ae_1 point would have to be used. This method, however, could not be reconciled with the method used thus far and would unnecessarily broaden the scope of this paper. The retarding parameter calculated from the two data points for the hypereutectoid steels is 1.57 and that from the first two points for the medium carbon series is 1.23. From the average value of 1.4 it may be concluded that chromium is comparable to manganese in its effect on pearlite.

The effect of nickel on the time for the beginning of pearlite formation at 1200 degrees Fahr. (650 degrees Cent.) taken from the data of Davenport (4) increases with concentration and is small up to nickel contents of 1.0 per cent. Up to this concentration the retarding parameter has an average value of about 0.55.

The various determinations of the retarding parameter made thus far in this paper are listed in Table II.

Effects of Manganese and of Molybdenum in Combination on the Time for Beginning of Pearlite Formation at 1200 Degrees Fahr. (650 Degrees Cent.)

In the steels used in the present investigation, neither carbon, manganese, nor molybdenum is fixed at any single concentration. However, there seems no reason to believe that the same insensitivity of the pearlite reaction rate to carbon content which was observed for the plain molybdenum steels would not also hold true for manganese-molybdenum steels. Therefore, in the analysis that follows carbon will not be considered as a variable for hypoeutectoid steels. The effects of manganese and molybdenum singly have been determined in the preceding sections. If their effects are quantitatively

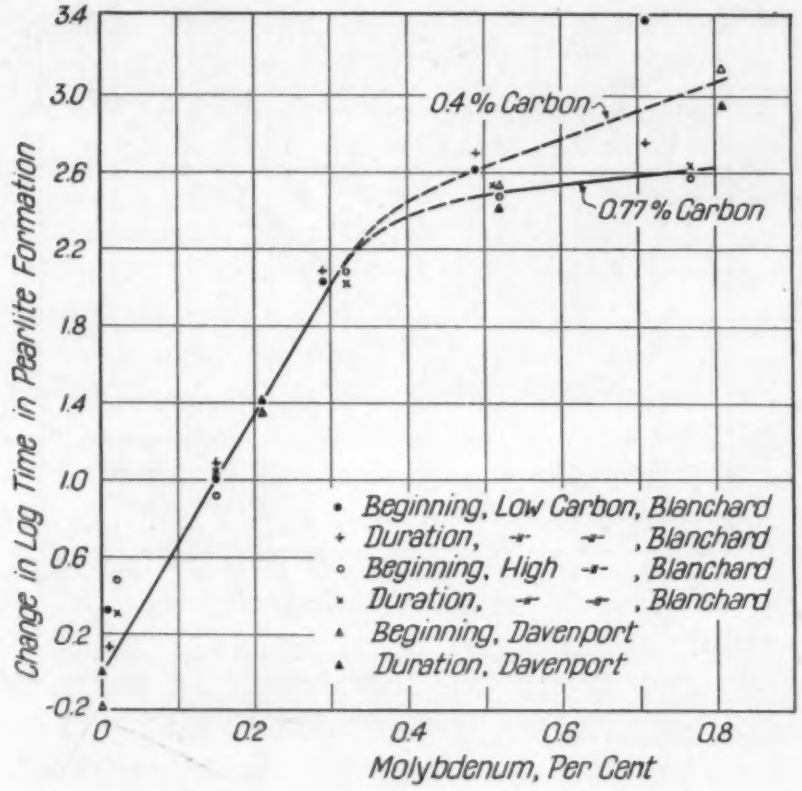

Fig. 15—Combined Isothermal Transformation Data of Figs. 10 to 12 for the Change in the Logarithm of the Pearlite Transformation and Nucleation Times as a Function of Molybdenum Content.

the same in manganese-molybdenum steels, then the slope of the log time-per cent manganese curve for a series of steels of equal molybdenum contents would be equal to the retarding parameter for manganese previously calculated in this paper from Davenport's data (4). Not all the steels are of a single molybdenum content, but several of them contain 0.45 $\pm$0.01 per cent. More data for the curve, however, may be obtained by correcting the log times for other steels to the values to be expected for 0.45 per cent molybdenum steels.

In order to obtain the best possible corrections for these molybdenum content differences, the four sets of data for the effect of molybdenum on the rate of pearlite formation (see Figs. 10 to 12) are combined in Fig. 15 to obtain one averaged curve to be used for correcting transformation times for different molybdenum contents. In adjusting the data to form this curve, the four curves of Figs. 10 to 12 were made to intersect at the 0.30 per cent molybdenum content (where data are most accurate) in order to fix the relative positions of all the data points on the log time axis. The intersection of the best straight line curve through the points at lower concentra-

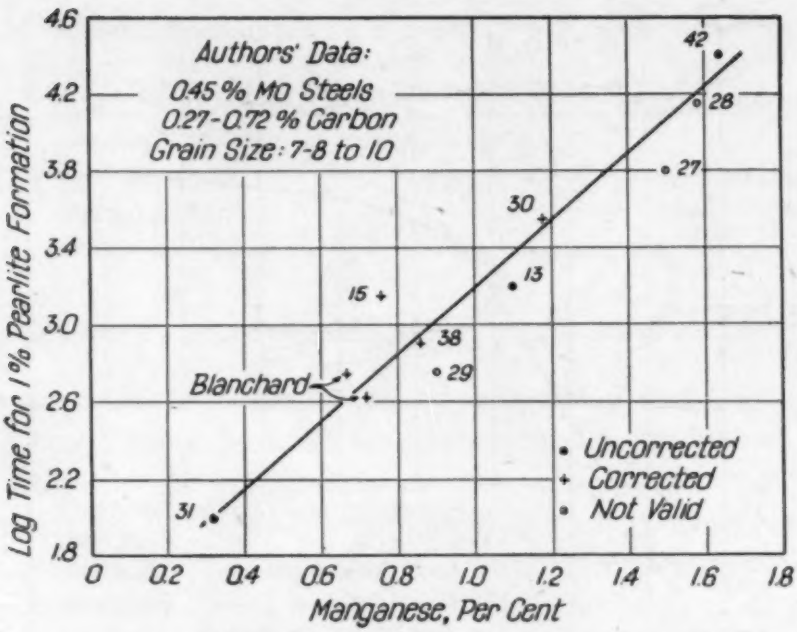

Fig. 16—Isothermal Transformation Data of the Present Authors for the Time for the Beginning of Pearlite Formation in Manganese-Molybdenum Steels at 1200 Degrees Fahr. as a Function of Manganese Content. Transformation Times are Corrected to a Constant Molybdenum Content of 0.45 Per Cent.

tions with the log time axis was then taken as the new zero point and thereby a curve was obtained for the change in log time as a function of molybdenum content.

The determination of the effect of manganese in the manganese-molybdenum steels studied in this paper is made in Fig. 16. Here the log of the time for the beginning of pearlite formation is plotted directly against manganese content for the 0.44 to 0.46 per cent molybdenum steels. In addition, the log time for steels of other molybdenum contents are plotted after correction to the values expected for a 0.45 per cent molybdenum content by the use of Fig. 15.

A straight line curve of slope 1.70 fits fairly well the experimental points of Fig. 16. In drawing this curve the encircled points were ignored as being somewhat inapplicable to the curve. In particular the transformation times at 1200 degrees Fahr. (650 degrees Cent.) for steels 27 and 28 were believed to be somewhat low because of the probable accelerating effect of the 0.20 and 0.16 per cent chromium present respectively, for the 1200 degree Fahr. (650 degree Cent.) level for these steels is on the flat portion of the pearlite nose. The point for steel 29 was ignored because this steel in contrast to all others showed no proeutectoid ferrite on transformation at 1200 degrees Fahr. (650 degrees Cent.) and was suspected of

being slightly hypereutectoid because it transformed more sluggishly when slightly decarburized. The corrected point for steel 15 is definitely far above the curve, perhaps because of the uncertainty in the correction for molybdenum differences at high concentrations. Points for two steels reported by Blanchard, Parke, and Herzig (2) , (3) are also plotted (only slight corrections for molybdenum being necessary), in order to show the agreement between these data and those of the former authors.

The value of 1.7 for the retarding parameter for manganese at 1200 degrees Fahr. (650 degrees Cent.) determined in Fig. 16 for 0.45 per cent molybdenum steels is very close to the probable "best" value of 1.6 determined in a previous section of this paper for plain carbon steels. Furthermore, the very good alignment of the curve with the two superposed data points of Blanchard, Parke, and Herzig (2), (3) shows that the effect of molybdenum in this series of steels is the same as in the 0.7 per cent manganese steels studied by the former authors. Therefore, it may be concluded that within the expected metallurgical uncertainty the retarding parameters of manganese and molybdenum for pearlite at 1200 degrees Fahr. (650 degrees Cent.) are the same in manganese-molybdenum steels as in simple steels, or in other words that the effects of molybdenum and manganese are the same when the elements are combined as when separate. The upper and lower limits for which this is true have not been determined.

It is not attempted to show here that the fairly constant values of the retarding parameter of a given element in the presence of different amounts of another element is fundamental in metallurgy but merely to present these data for manganese-molybdenum steels as an aid to heat treatment. It is hoped, however, that more research of this same type will be initiated.

Limiting Cooling Velocities for the Suppression of Pearlite Calculated from Chemical Composition

Data have now been obtained for the effects of carbon, manganese, and molybdenum on the time for the beginning of pearlite formation at the pearlite nose, assuming that the nose is maintained at 1200 degrees Fahr. (650 degrees Cent.). Numerical values may now be substituted in the "reference" formula of Grange and Kiefer (1) for the limiting cooling rate for pearlite suppression. Thus

$$R = \frac{170}{1.5 \times t_o \times 10^{1.6 \times (\%\ Mn)} \times 10^{6.7 \times (\%\ Mo)}} \qquad \text{Equation 5}$$

where the values of 1.6 and 6.7 are used as the most probable values of the retarding parameter of manganese and molybdenum, and where t_o is the extrapolated value of the time for the beginning of pearlite formation in a steel containing no manganese and no molybdenum. This value can be obtained with best accuracy from the data of Blanchard, Parke, and Herzig (2), (3) shown in Fig. 11. Here the log of the time for beginning of pearlite formation is 1.16 for a hypoeutectoid steel of 0.70 per cent monganese and 0.15 per cent molybdenum. Therefore

$$\log t_o = 1.16 - 0.70 \times 1.6 - 0.15 \times 6.7 \qquad \text{Equation 6}$$

or $t_o = 0.107$ seconds

Equation 5 may therefore be rewritten in a form more suitable for use as

$$\log R = \log 170 - \log 1.5 - \log t_o - 1.6 \times (\%\ Mn) - 6.7 \times (\%\ Mo) \qquad \text{Equation 7}$$

$$\text{or } \log R = 3.02 - 1.6 \times (\%\ Mn - 6.7 \times (\%\ Mo) \qquad \text{Equation 8}$$

For example, in a steel of 1.0 per cent manganese and 0.3 per cent molybdenum,

$$\log R = -0.53$$

or $R = 0.30°$ F./sec. $= 18°$ F./min.

The above formulas are approximations to the reference formula of Grange and Kiefer (1) made by assuming that the pearlite nose in a manganese-molybdenum steel is at 1200 degrees Fahr. (650 degrees Cent.), that the grain size is No. 8, that carbides are in solution before cooling, and that the residual alloys are low. The assumption that the pearlite nose be at 1200 degrees Fahr. (650 degrees Cent.) requires some explanation. It is observed that the retarding parameter for manganese is greater than 1.6 above 1200 degrees Fahr. and smaller below this temperature. Therefore, a lowered pearlite nose produced by a relatively high manganese content would occur at a shorter time than that assumed in equation 8, and the rate calculated by the use of this equation would be smaller than that calculated directly from the Grange and Kiefer formula using actual S-curve data.

Limiting Cooling Velocities Compared with Total Alloy Content

The value of the limiting cooling velocity as calculated by Grange and Kiefer's formula from direct S-curve data cannot yet be corre-

lated with chemical composition except where equation 8 can be used. However, with the data obtained in foregoing sections it is possible to evaluate the total alloy content of a manganese-molybdenum steel in terms of equivalent manganese content without ignoring the presence of residual elements.

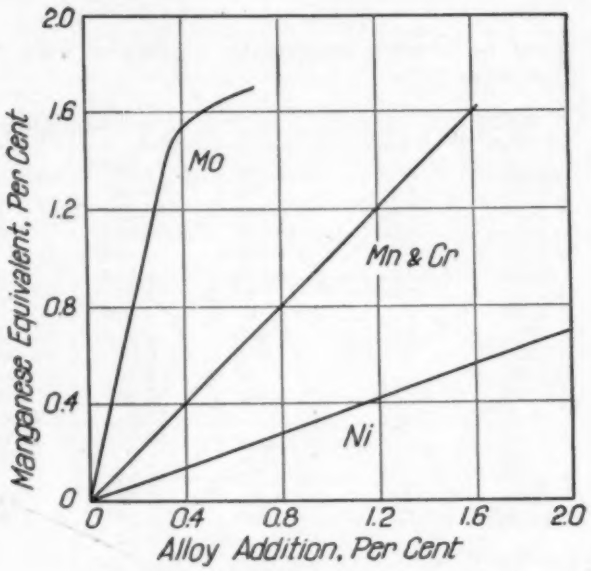

Fig. 17—Relation Between the Manganese Equivalent of an Alloy Addition in Retarding Pearlite Formation at 1200 Degrees Fahr. and the Actual Alloy Addition.

The retarding parameters of molybdenum and manganese at 1200 degrees Fahr. (650 degrees Cent.) are in the ratio 6.7 : 1.6. Therefore, molybdenum is about 6.7/1.6 or about 4.2 times as effective in retarding pearlite at this temperature as an equivalent amount of manganese, for it would require about 4.2 times as much manganese to produce the same effect as a given amount of molybdenum. Hence the total alloy content of a manganese-molybdenum steel may be represented as its manganese content plus 4.2 times its molybdenum content, provided the molybdenum content is not above 0.3 per cent. The effectiveness of higher molybdenum concentrations may be determined by using Fig. 15. The effectiveness of residual nickel may be determined from the value of its retarding parameter relative to that of manganese. The effectiveness of residual chromium as stated before may be considered comparable to that of manganese, and that of silicon may be considered negligible (see the Atlas of Isothermal Transformation Diagrams, United States Steel Corporation) (7). The manganese equivalents of various amounts of molybdenum, chromium, and nickel calculated as described above are plotted in Fig. 17.

In Table III are tabulated the limiting cooling velocities calculated from the S-curves of this paper and from some other isothermal transformation diagrams according to the formula of Grange and Kiefer, together with the necessary data used in the calculations.

Table III
Limiting Cooling Rates for Pearlite Suppression Calculated from S-Curve Data

Steel	Ae_1 (° F.)	T_{pn} (° F.)	Log t_{pn}	t_{pn} (sec.)	Limiting Cooling Rate (° F./sec.)	(° F./min.)	Total Alloy Cont. (% Mn)
13	1380†	1200	3.25	1,800	0.067	4.0	2.68
15	1370†	1200	3.4	2,500	0.045	2.7	2.44
27	1335†	1175	3.65	4,500	0.024	1.44	3.38
28	1380†	1175	4.0	10,000	0.014	0.82	3.30
29	1340†	1175	2.8	600	0.183	11.0	2.63
30	1340†	1175	3.0	1,000	0.11	6.6	2.63
31	1340†	1100	1.8	60	2.7	162.0	2.03
42	1340†	1150	3.8	6,500	0.020	1.17	3.27
38	1370†	1200	3.1	1,500	0.076	4.5	2.61
A*	1350	1200	...	160	0.63	38.0	1.90
B*	1350	1200	...	500	0.20	12.0	2.34
C*	1350	1200	...	700	0.14	8.4	2.48
SAE 5140§	1360	1175	...	4.8	26.0	1600	1.64
SAE 3140§	1305	1100	...	6.0	23.0	1400	1.67
SAE 4140§	1365	1200	...	50	2.2	130	2.53
SAE 4340§	1300	1175	...	2700	0.031	1.8	3.57
NE 8744‡	1315	1200	...	110	0.70	42.0	2.52
NE 8949‡	1310	1175	...	600	0.15	9.0	3.32

*Blanchard, Parke, and Herzig (2).
§Davenport (4).
‡Atlas of Isothermal Transformation Diagrams (7).
†Ac_1 point.

Rates for some steels outside the manganese-molybdenum classification are included.

These calculated rates are plotted in Fig. 18 on logarithmic scale against total alloy content in terms of the manganese equivalent determined by the use of Fig. 17. Superposed on the plot is a straight line curve representing the theoretical values given by equation 8.

The points for steels 13, 30, 38, A, B, and C fall close to this curve. Those for steels 31, 27, 28, and 42 are above this theoretical curve probably because the pearlite nose in these steels is below 1200 degrees Fahr. where the effectiveness of manganese is less strong. As in Fig. 16, the points for steels 15 and 29 do not correspond with the other data.

The data points for steels containing significant amounts of chromium occur at higher calculated rates than those for manganese-molybdenum steels of equivalent alloy content. Unless the manganese equivalent of chromium has been overestimated, this fact indicates that alloy content is not fully effective in retarding pearlite when chromium is present. Whether this deficiency in chromium

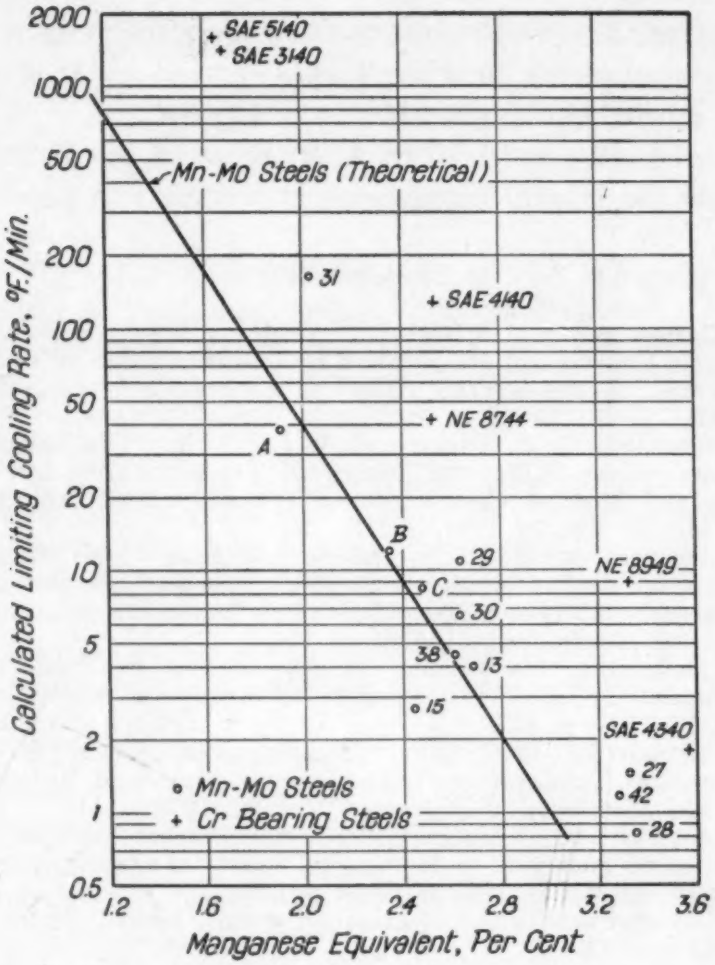

Fig. 18—Data Plot of the Values of the Limiting Cooling Rate for Pearlite Suppression Calculated from S-curve Data According to the Formula of Grange and Kiefer (1) in Relation to Total Alloy Content as Determined from Fig. 17.

steels is due to the inherent behavior of the steel under optimum conditions or to incomplete solution of carbides under ordinary conditions does not affect the conclusion that pearlite may be more easily retarded in a plain manganese-molybdenum steel.

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PRACTICAL APPLICATION OF STATISTICAL METHODS IN A QUALITY CONTROL PROGRAM

By W. T. Rogers

Abstract

This paper presents examples of the application of statistical methods in a quality control program. Three general methods of handling routine data are discussed: frequency distributions, control charts, and correlation, with a number of actual examples demonstrating each application.

The advantages of the control chart method of presenting data are compared to those of the frequency distribution in both routine and experimental problems, and the inadequacy of the frequency distribution is pointed out. Problems in simple and multiple correlation, taken from actual experience, are presented, with a final example showing the results of a co-ordination of correlation and control charts.

An effort has been made to keep all discussion practical, although it is presupposed that the reader has an elementary knowledge of statistics.

IN the past twenty years, and particularly since the publication of Dr. W. A. Shewhart's book in 1931 (1),[1] industry in general and the steel plants in particular have been gradually adopting statistical methods as a practical means of obtaining quality and production control. In a few plants these methods have been developed into a regular adjunct of the department responsible for the control of quality and are developed to the extent that they are the guide of plant management in following the level of quality being produced. In other plants their use is still in the experimental stage and has only recently been given impetus by the need for obtaining a maximum output of good material. As far back as 1933, attention was being called to the useful application of these methods by steel plant metallurgists, as evidenced by W. C. Chancellor's (2) paper of that year and a more recent publication in 1937 by H. J. Hand (3). In

[1]The figures appearing in parentheses pertain to the references appended to this paper.

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these reports and in subsequent discussions, both written and oral, the practical application of this new technique has been ably demonstrated. Up to 1939, however, these applications have been confined to a very limited field and have been applied only in isolated cases, principally to experimental or short range projects. There is no evidence up to this time of their extended use as a matter of routine procedure.

With the advent of the war in 1939 and the rapidly increasing demand for more and more good product in the shortest possible time, engineers have been increasingly anxious to use every means at their disposal not only to produce more but to obtain an increasingly larger percentage of good product. This increase in the proportion of good product can be obtained only by maintaining a high and consistent level of quality in all related operations, and it is not surprising to persons well versed in the technique of quality control by statistical methods to find that their services are now in great demand. The statistician dealing with quality control in a manufacturing plant of practically any description finds, almost before he can realize the fact, that he has changed from the status of statistician to that of quality control engineer. In fact, the need for this type of personnel has developed so rapidly that the War Production Board, in conjunction with the American Standards Association, has recently put on a program of intensive short term courses in the methods of applying statistical technique to quality control. These methods have, in this manner, been demonstrated not only to the individual engineer, but to the executive branches of plant management directly concerned with production and quality.

In consequence of the recent development which has taken place in this field, it is felt that an effort should be made by those using these methods to make the practical results obtained available to others. Most discussions on the subject of the maintenance of quality control by statistical methods deal with some one particular application or one phase of a control program, and it was not possible to get a picture of an over-all program in any instance. It is for this reason that the following commentary on "The Practical Application of Statistical Methods in a Quality Control Program" has been prepared, and it is hoped that the discussion presented will stimulate others in the same field to publicize other programs to the end that the better methods may be adopted by all who are engaged in this type of work.

The scope of the material presented includes a comparison of two common methods of presenting data together with certain representative practical applications which, it is believed, will furnish a clear and concise picture of the use of statistical methods in connection with a quality control program. The subjects discussed are as follows:

1. Frequency Distributions.
2. The Advantages of the Control Chart Over the Frequency Distribution.
3. The Frequency Distribution as Compared to the Control Chart in Evaluating Experimental Results.
4. Examples of Simple and Multiple Correlation and the Place of Correlation in a Quality Control Program.
5. The Combined Use of Control Charts and Correlation in a Quality Control Program.

It is believed that the subsequent discussion of these five subjects will carry those interested through a complete application of this procedure and point out the advantages to be obtained from the use of the methods involved.

Inasmuch as the material herein presented is entirely objective in character, all methods of technique, such as the construction of frequency distributions, the calculation of control limits, methods of correlation, curve fitting and other statistical adjustment of data, are purposely avoided. The methods used are well established and can be found in the bibliography.

Frequency Distributions

In any industry producing material by repetitive processes, constancy of the factors involved is not obtained. The numerous variables associated with quality and production occur with more or less mathematical regularity about any one desired value. Generally, the desired value is the most frequently encountered, with values on either side occurring with less and less frequency between certain limits which can, in most cases, be determined by statistical methods. It is, therefore, necessary for anyone, either directly or indirectly associated with quality and production control, to have some conception of the frequency distributions with which he is likely to come in contact, and also to be able to evaluate the information which can be obtained from these distributions in terms of its practical value and application. Distributions are of many types and forms, character-

ized by certain mathematical values, such as the average or central tendency, the standard deviation or extent of variation, the skewness or symmetry and the flatness or the relative occurrence of other values about the average. As much as it is desired to have normal distributions which will have an average ($\overline{X}$) at the median or mid-point between the two extreme values, a range of values such that the average plus or minus three times the standard deviation ($\overline{X} \pm 3\sigma$) includes all the variation in observed values, a skewness of zero indicating perfect symmetry and a flatness of three indicating normal occurrence of distributed values with respect to each other, this is the least likely form in which they will occur. Regardless of the characteristic shape of any curve, there is one thing which should be known about any distribution, and that is whether or not it takes its shape from a chance or nonchance system of causes. To determine this fact correctly is one of the important tasks of the quality control statistician.

Chance causes may differ from time to time so that a distribution of any one variable will have a different form in one period than in another. As time progresses, an element may periodically enter and leave the system. A factor may appear in different magnitudes from time to time, depending on previous processes, or a completely new factor may enter or an old established factor leave the system. Regardless of these situations, one thing is constant whether statistical control is practiced or not. In every manufacturing process, whether it be the canning of tomatoes or the building of a B-29 airplane, the manufacturer aims for a definite mark in controlling all of his processing variables to the end that the most satisfactory product may be made with the greatest economy. A quality control program then must have as its aim the adoption of the best available methods of determining the state of control and presenting the information in a manner which will point out to those concerned any deviation from normal variation. With this in mind, the merits of the frequency distribution as a means of furnishing information will be discussed.

A Distribution From a Constant System of Chance Causes

In Fig. 1-A is presented the frequency distribution of numbers obtained by casting a pair of dice 480 times and recording the frequency of recurring numbers. This distribution is for the purpose of

illustration, as will be subsequently developed. For the present, however, it can be assumed that it was obtained by performing the 480 operations in the same manner and with unbiased dice. On the basis of this assumption, the resultant numbers should occur in a predetermined probability distribution with the following relative frequency:

Number	Per Cent Occurrence
2	2.78
3	5.56
4	8.33
5	11.11
6	13.89
7	16.67
8	13.89
9	11.11
10	8.33
11	5.56
12	2.78

Provided the dice are not biased and the operation of casting is constant, this is the type of distribution which will always be obtained, and in any subsequent repetition of the experiment the results will check within limits based on the laws of probability. In other words, knowing the chance system of causes from which a distribution arises, it is possible by statistical tests to arrive at a conclusion as to whether or not a group of observations can be regarded as having originated from the known system. The one question which must be answered when analyzing a frequency distribution is, "Do these data come from a constant system of chance causes?" From conventional tests applied to the distribution given in Fig. 1-A, it would be assumed that it does originate from a chance system. Inasmuch as one of the first procedures in analyzing data is the construction of a frequency distribution, it is obvious that an evaluation of the information obtained by this procedure must be made. With this in mind, the remaining five curves in Fig. 1 are presented.

Typical Distributions Actually Obtained in Practice (Mill Yield)

In every process the yield of finished material is vital to the economic manufacturing of a product. In the production of pipe, it is therefore necessary to know the variations in yield from day to day and to strive constantly for yields as high as possible, consistent with

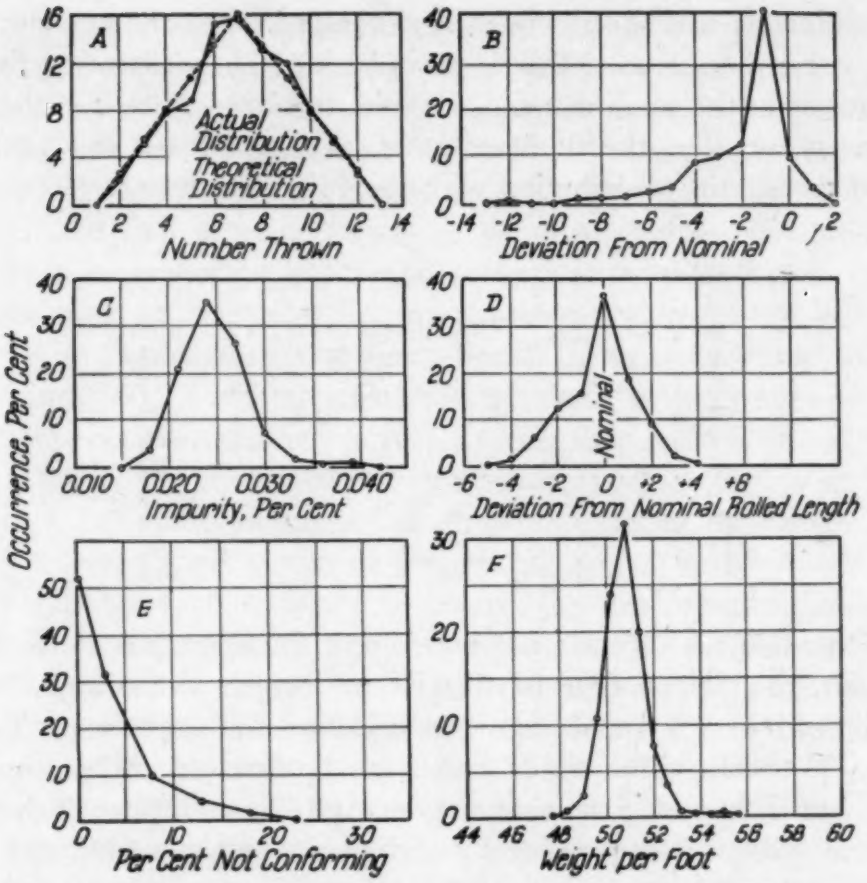


Fig. 1—Examples of Routine Data Presented as Frequency Distributions. (A) Frequency distribution of the occurrence of numbers when throwing the dice 480 times. (B) Frequency distribution of mill yield. (C) Frequency distribution of per cent impurity in a raw material. (D) Distribution of deviation from nominal rolled billet length from a bar mill. (E) Frequency distribution of per cent of nozzle assemblies not giving satisfactory results. (F) Frequency distribution of weight per foot characteristic of finished product. (In Fig. 1A the solid line is "actual distribution" and the dotted line "theoretical distribution".)

the maintenance of high quality. Some losses are more or less constant, such as scaling due to heating, irregular ends (crop ends) which must be discarded, and other somewhat erratic losses due to processing difficulties. Each mill makes various sizes and classes of pipe, all having a theoretical standard yield, the deviations from which are shown in Fig. 1-B. Note how this curve is skewed to the right, indicating that close to standard yield is being obtained most of the time and that deviations from this standard are more frequently encountered on the negative side.

Impurity in a Raw Material

Certain impurities in raw materials detract from the processing

quality in the steel plant when present in more than nominal amounts and must be controlled at relatively low levels in the finished product. A distribution of one such element is shown in Fig. 1-C, illustrating a curve skewed to the left, indicating that in most cases the concentration of this constituent is at a relatively low level.

Billet Length From a Bar Mill

The length of billets from a bar mill shown in Fig. 1-D has a definite effect on the size control in subsequent operations and must be controlled within narrow limits. The distribution presented shows a very close range of values (each unit is ¼ inch), and apparently good control is present.

Material Not Conforming to Standard Service Requirements

There are many cases in industry where it is necessary to evaluate a condition by the method of attributes instead of variables. In other words, instead of evaluating a condition numerically, it must be measured by a certain characteristic to which it either does or does not conform. An example of this type is the performance of the nozzle assembly on an open-hearth pouring ladle. In order to facilitate the pouring operation, it is necessary to have a nozzle assembly which will operate smoothly and give a good shut-off from ingot to ingot, and inasmuch as the evaluation of this condition depends on the judgment of the operator, it is evident that it can only be recorded as satisfactory or unsatisfactory. A frequency distribution of the daily percentage of nozzle assemblies giving unsatisfactory results is presented in Fig. 1-E. This curve is the familiar fraction defective curve found in practically all cases where conditions are considered as conforming or nonconforming.

Weight Per Foot of Finished Pipe

Weight per foot of finished pipe has rigid acceptance specifications and must be controlled within very narrow limits. A distribution of this variable on one lot of pipe is shown in Fig. 1-F where it is evident that a very confined distribution without undue skewness is obtained.

Distributions of the types shown, and many others, are con-

stantly coming up in the course of routine analysis of data. While it is possible by statistical tests to determine whether certain curves are normal and others not, it is impossible from a frequency distribution alone to arrive at an assumption as to the shape of the parent distribution from which any data are obtained. All assumptions relative to the parent distribution must be predicated on the immediate distribution at hand, and there is no assurance that this is representative of day-in, day-out performance. At best, the frequency distribution shows only the relative occurrence of individual items during one period of time, and in no case, even though a distribution proves by statistical tests to be normal, can it be assumed that the cause system normally operating is that represented by the data so grouped. Neither can we say that data in this form tell anything about the state of statistical control prevailing during the period in which they were collected. It is, therefore, evident that if this kind of information is desired, and in a quality control program it is not only desired but necessary, it must be obtained by some other method. The best method advanced to date for doing this is the control chart.

The Advantages of the Control Chart Method Over the Frequency Distribution

In the foregoing discussion, six frequency distributions have been presented, about which it has been attempted to draw some conclusions concerning the population or universe from which they were derived and to obtain some knowledge as to whether the occurrence and recurrence of values observed is due to chance or nonchance causes. Viewing these curves from a negative angle, it is obvious that they tell us little of value about the homogeneity of the data presented. It is not possible to determine definitely whether the different variables are in any state of statistical control; that is, whether or not they are fluctuating within the limits of probability. It may be inferred that nonchance causes do not exist, as in Fig. 1-A, or that a distribution lacks control, as in Fig. 1-B. Even though statistical tests bear out the inference of visual inspection and considered judgment, one can only arrive at the conclusion that it is *quite probable* that a given set of data did or did not come from a controlled system of chance causes. Furthermore, it cannot be determined from a frequency distribution whether more than one level of control is included in the data, nor can the existence of trends be detected.

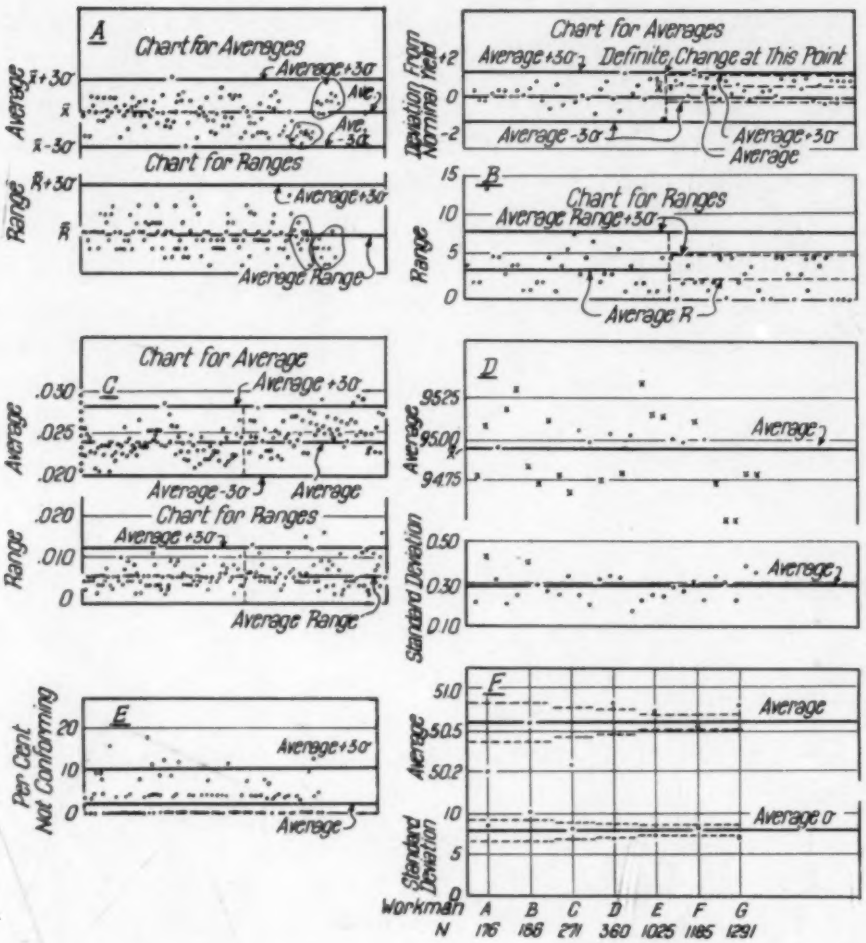


Fig. 2—Examples of the Control Chart Method of Presenting Routine Data. (A) Control chart for averages and ranges of groups of four occurrences of numbers when throwing two dice. (B) Control chart for mill yield average and range of groups of four values. (C) Control charts for average and range of per cent of impurity in a raw material. (D) Control chart for average and standard deviation of billet length from a bar mill. (E) Control chart for per cent of nozzle assemblies not giving satisfactory results. (F) Control chart for weight per foot characteristic of finished product.

One is therefore forced to the conclusion that frequency distributions are not suited to the routine analysis of observed data in a manufacturing process. The variables presented in Fig. 1 are compared in each case to the control chart method shown in Fig. 2 in answer to the question of "How then shall data be collected so as to furnish the maximum amount of information?"

In the case of the frequency distribution shown in Fig. 1-A, wherein it was assumed that the data occurred in a normal chance manner, the control chart (Fig. 2-A) for averages and ranges of groups of four numbers plotted in chronological order tells a different story. It can be seen here that points plotted are well controlled

up to the last 20 values, where it is evident from the grouping that some extraneous factor was introduced, first to make 10 low averages and then 10 high averages. The range of values within each group of four, as shown by the last 20 points, is also definitely lower. The reason for this lack of control was the elimination of throws which presented a 5 or a 6 on 40 consecutive throws, and the elimination of throws which presented either a 1 or a 2, thereby forcing 10 consecutive averages down and the succeeding 10 up, thus lowering the range on all of the 20 groups. This biasing of the data, although not apparent in the frequency distribution, is readily discernible by the control chart.

In Fig. 2-B, which presents a control chart for deviation from nominal mill yield, the same data as used in Fig. 1-B are shown chronologically. It will be noted in this chart that the skewness shown in Fig. 1-B is not due to lack of control, but rather to two different levels. The last half of the chart shows that at a certain point a definite change took place which raised the average and also narrowed the chance limits of variability. Both sections show that control was maintained all the time, the last half representing the elimination of *chance factors* which were present in the first half.

Fig. 2-C is the corresponding control chart for Fig. 1-C, the frequency distribution of an impurity in a raw material. The frequency curve shows a fairly uniform distribution, slightly skewed to the left, which would probably be assumed to arise from a chance system of causes, but the control chart again gives a different story. Here it is evident from the chart that the first part of the data is cyclic in nature, and the last half shows a definite trend toward the high side. In the chart for range, it is evident that in addition to the averages being high in the latter half, the individual values are also more erratic in their occurrence.

Fig. 2-D, showing a control chart for billet length from a bar mill, is a reconstruction of the data in Fig. 1-D, with the exception that in Fig. 1-D the length is plotted as deviation from nominal; whereas in the control chart, points are plotted as actual averages. Here again is seen the advantage of the control chart in presenting a complete story of the occurrence of observed values. Looking only at the frequency distribution, it will be noted that it lies between nominal plus or minus 1 inch. This may be considered precision work on a heavy mill of the type involved, but a study of the control chart reveals an entirely different condition. Here it is evident that

the chart for averages shows no control whatsoever from group to group, while the chart for standard deviations shows relatively good control. From this presentation it is deduced that control within each group of points is good, but there are assignable causes of variation from group to group. Limit lines are not shown on this chart, due to nonuniform sample size, and standard deviation is used instead of the range, due to the number of items in each group.

Fig. 2-E, a control chart for percentage of items not giving satisfactory results, shows a lack of control which would not be suspected from the frequency distribution shown in Fig. 1-E.

In Fig. 2-F, the data for weight per foot shown in Fig. 1-F are broken down into averages and ranges by workmen weighing the material. Here it is evident that even though a very confined total distribution is obtained, there is a significant difference between each individual weighman.

It is readily apparent from the comparisons shown that if one wishes to know the whole story relative to the chance or nonchance occurrence of observed values, the control chart is of inestimable value, in that it tells:

1. The state of statistical control.
2. The evidence of more than one level of control.
3. The existence of nonchance causes.
4. The existence of cycles or trends.

In addition to the above reasons, the utility of the control chart is further enhanced by the fact that it affords the user a tool which will quickly point out changes due to a shift in the cause system responsible for fluctuations in a given variable.

The Frequency Distribution Compared to the Control Chart in Evaluating Experimental Data

It is often necessary, in an industrial process, to evaluate the results obtained by changing a process in order to determine whether an additional change will give a more uniform product. A common method of making such comparisons is the frequency distribution from which an opinion is formulated on the basis of the congruity of two curves. Two such comparisons are made in Figs. 3-A and 3-B. In 3-A two methods of controlling the billet length on a bar mill are compared and by this comparison, it is obvious that no definite conclusion can be drawn from the curves shown. It would probably be

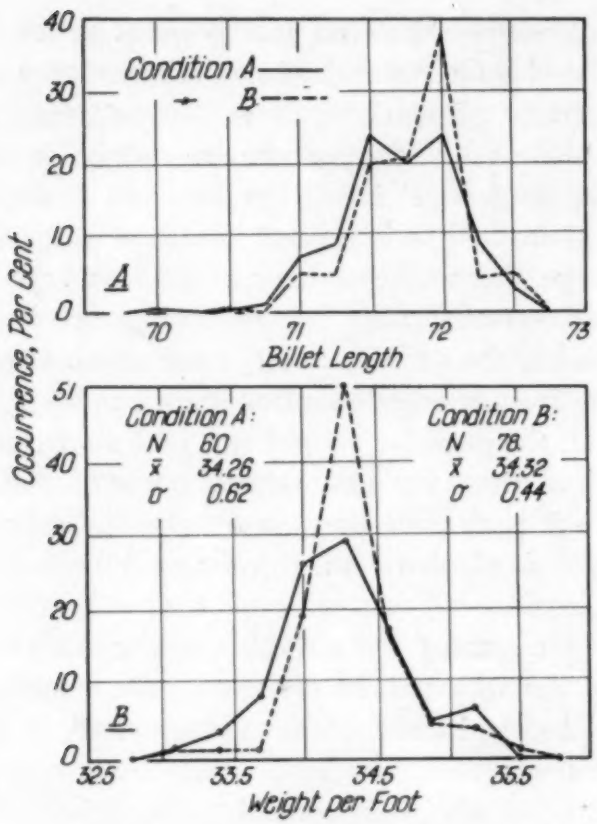


Fig. 3—Frequency Distribution Method of Presenting Experimental Data. (A) Frequency distribution comparing two methods of controlling billet length on a bar mill. (B) Frequency distributions of weight per foot of a seamless pipe product. Comparison of two methods of production.

assumed by this method of approach that there is no difference in the results obtained or at best that the results are inconclusive. Fig. 3-B shows the results obtained by two methods of controlling the weight per foot of a seamless pipe product. Here the conclusion would be made that method B is superior to method A in that a definitely narrower distribution is produced.

By the control chart method, however, as shown in Fig. 4, a very much more conclusive picture of the actual results is presented. In this case it may be seen that, in the case of average billet length, there are three points outside the limits of chance variation with one condition and four with the other as indicated by the points plotted with an X (limits being variable due to nonuniform lot size) with no apparent difference in the remaining averages with either condition. A further examination with reference to standard deviations reveals that the condition to the right of the figure has a lower level of control, indicating a more confined distribution within each lot.

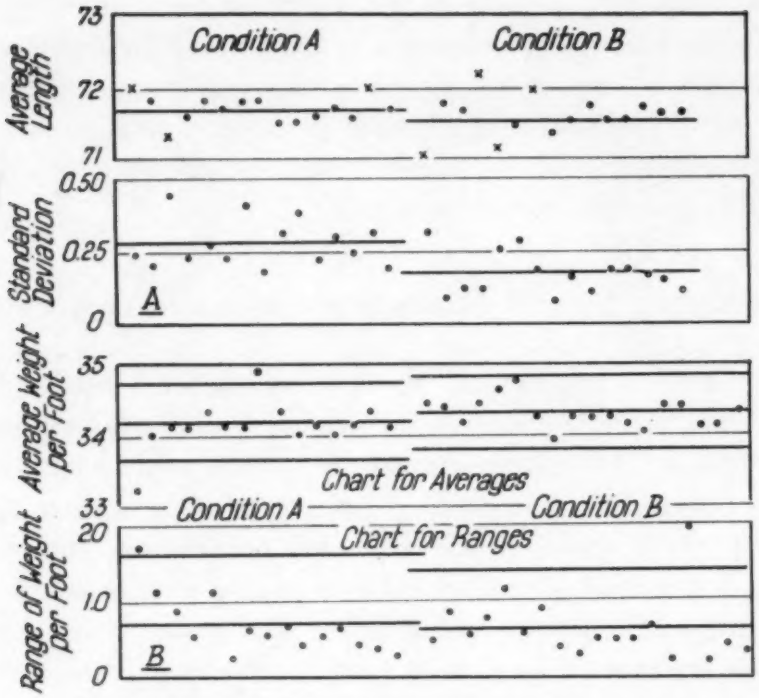


Fig. 4—Control Chart Method of Presenting Experimental Data. (A) Control chart method of comparing two methods of controlling rolled billet length on a bar mill. (B) Control chart method of comparing weight per foot on two methods of production of seamless pipe.

It is evident then that whereas a comparison of frequency distributions gave no information on which conclusions could be based, the control charts throw a considerable light on the results and enable definite conclusions to be drawn.

The control chart method applied to the data included in the frequency distributions of Fig. 3-B is shown in Fig. 4-B. The frequency distributions led to the assumption that condition B was superior to condition A due to more confined distribution. Here it will be seen that this conclusion would be erroneous. Upon examination of the chart for averages, it is evident that the wider spread obtained in the frequency distribution of condition A is due to two points representing eight individual values which are outside the limits of chance variation. Aside from these points, control is better with condition A. Condition B shows that, although all points are within the calculated control limits, two definite levels were obtained, the first six points showing a different average than the remaining points. The charts for ranges show that, in both cases, results were erratic at the beginning of each test and increasingly uniform as the test progressed.

It is, therefore, demonstrated by these two examples in dealing with experimental comparisons in the line of production, wherein chance factors of variation are numerous and nonchance conditions likely to arise, that the control chart is more desirable than the frequency distribution. It is believed that, in general, this conclusion will hold true in this type of analysis.

Examples of Simple Correlation in a Control Program

Control of a production process consists essentially in determining immediately the degree of success being attained and a system of repeating this determination from time to time as a matter of routine. There are, however, certain processes where the actual measure of effectiveness is not only time consuming, but in addition may be so destructive of material that repeated measurements at short intervals are economically undesirable. An instance of this kind is the metallic coating of steel pipe by the hot dip galvanizing process. In this operation when it is desired to know exactly the coating weight being applied, it is necessary first to select a piece of pipe from the production line which is felt to be representative of the material being produced. Representative test pieces must then be cut from various sections of the pipe, which are sent to a laboratory to be carefully weighed, stripped of their zinc coating and re-weighed. Coating weight can thus be determined by difference. It can readily be seen that while this method is of value to those concerned with the operation, it is time consuming and loses much of its effectiveness as a method of control, in addition to being a "destructive test" method.

In cases of this kind, it is generally much more desirable to have a method which will give rapidly, without the destruction of finished product, a measure which indicates a certain value that is highly correlated to the variable under consideration. There are instruments now available which give quickly, and without destructive testing, numerical values which are highly correlated to the actual coating weight. It is necessary only to obtain a suitable instrument and to develop the relationship between the values obtained by its use and the actual coating weights obtained by more precise methods. This is shown in Fig. 5-A, and from this statistical correlation the control of the process can be checked at short, equally spaced intervals with the assurance that actual results are within whatever probability limits desired. Thus, a rapid and efficient means of control is effected.

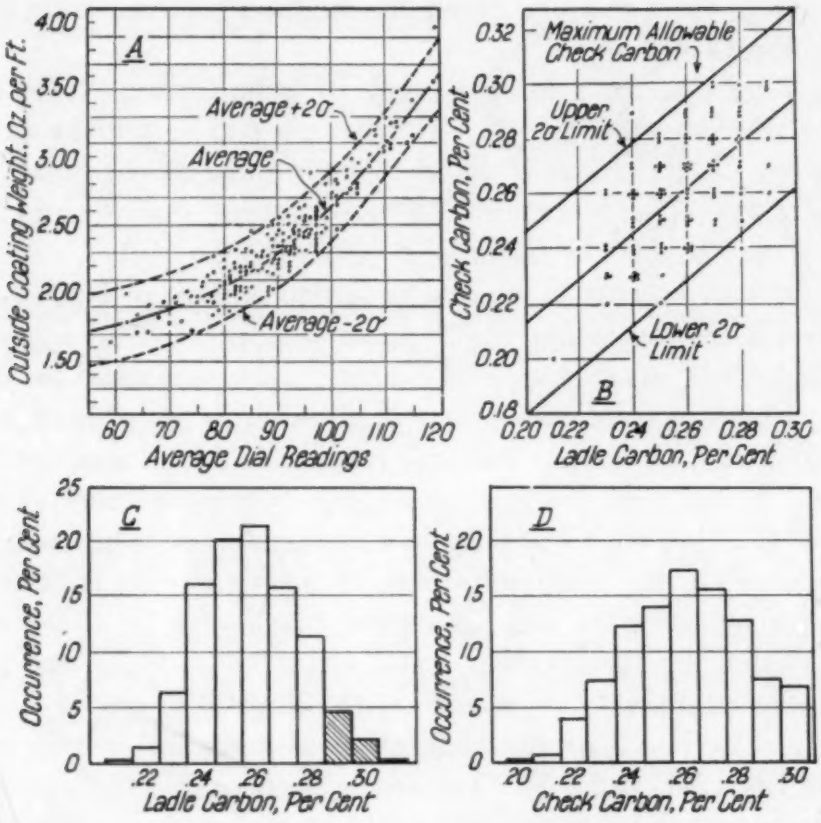


Fig. 5—Examples of the Use of Simple Correlation in a Control Program. (A) Relation between instrument dial reading and coating weight. (B) Relation between ladle carbon and check carbon. (C) Frequency histogram of per cent ladle carbon obtained in one grade of steel. (D) Frequency histogram of per cent check carbon obtained in one grade of steel.

Another type of problem frequently encountered in industry is one in which the end result of a process must be determined in one of the preliminary stages of manufacture. A good example of this is the chemical analysis of steel. It is familiar to all who are connected with steel plant metallurgy that in most cases steel is bought and sold, in so far as chemistry is concerned, on the basis of chemical analysis of the finished product (check analysis). The steel manufacturer, however, must accept steel based on the analysis of the metal in the as-cast condition (ladle analysis). The ladle analysis which is taken directly from the molten steel represents the greatest degree of homogeneity in chemical constituents that is possible to obtain while the check analysis, due to the physical characteristics of the solidification process, is subject to considerable variation. In view of this condition, it is necessary for the metallurgist, charged with the responsibility of accepting material to meet certain specifi-

cations, to know something about the relation between ladle analysis and check analysis.

This relation for a limited range of ladle carbons from 0.21 to 0.29 per cent is shown in Fig. 5-B. Here it will be noted that on the average the check carbon is practically the same as the ladle but for any one value of ladle carbon it can be expected that the check analysis may vary by plus or minus 0.03 per cent (2σ limits). Assuming that a customer specifies a maximum carbon content of 0.304 per cent and physical requirements which make it imperative to apply heats with the highest possible carbon content, a decision must be reached as to how high in ladle carbon it is permissible to go without running too great a risk of having material fabricated which will not be applicable. A study of the relation in Fig. 5-B shows that at a ladle carbon of 0.27 per cent it is relatively certain that no values will be obtained exceeding the maximum of 0.304 per cent. At 0.28 per cent the chance increases to such an extent that it can be expected that values outside the specification limits might be obtained and at 0.29 per cent the probability of exceeding the limit is practically prohibitive (4). The result of the application of selection on this basis is shown in Figs. 5-C and 5-D. The shaded part of Fig. 5-C (ladle carbon) represents the heats rejected as not applicable on the basis of the relation shown in 5-B and the check analyses obtained were as presented in Fig. 5-D.

Thus it can be seen that by setting up statistical relationships based on reliable past experience, statistical methods of control are not only practical but are a valuable adjunct to any control program.

Multiple Correlation[2]

The place of multiple correlation in a quality control program has only recently been given adequate recognition. In the past, when trying to isolate the individual effect of numerous variables on a quality characteristic it was felt necessary to accumulate an enormous quantity of data so that each supposedly related variable could be more or less stabilized. It was assumed that for each unit of variation in the one factor, other factors would be held constant due to the large number of items. While no serious fault is to be found with this method, since analysis made in this manner can be compared to scientific engineering knowledge and a determination made as to

[2]For Methods of Correlation Analysis, see reference (5).

whether results obtained are logical and valid, it has the condition imposed that the original quantity of data must be excessively large and in ordinary quality control work it must be ruled out on this basis. Fortunately, however, research workers in other fields have developed methods of multiple correlation whereby regression constants may be worked out simultaneously on several variables at one time. With the adoption of these methods in industry it has been found that highly reliable results can be obtained with a limited supply of data as is shown by the following example:

It was desired to determine the factors affecting the surface quality of a certain grade of open-hearth steel billets with the view to reducing the inspection rejects and thereby increase the yield of good product. A review of the data revealed that there were 131 heats on which complete information was available and there were 21 variables which were thought to be contributing causes to the variability of surface quality. Of these 21 factors the following 6 were considered of sufficient significance to be included in a final multiple correlation analysis:

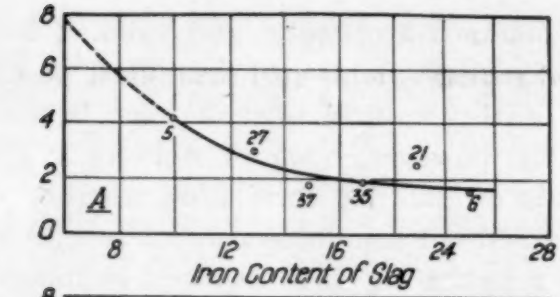
1. Iron Content of Open-Hearth Slag.
2. Casting Temperature.
3. Time in Pits (Ingot Reheating Time).
4. Mold Temperature.
5. Mold Wash.
6. Nozzle Condition.

The relations developed are as shown in Fig. 6 and based on the results of this analysis it was possible to reduce the index of rejections from 2.6 to 0.4.

Handling of problems similar to this is the regular function of a quality control department and such studies should be continuous for it is only by recognizing and controlling quality related factors that permanent improvements can be achieved. The value of whatever control charts are maintained will be greatly enhanced if it is possible to point out why an observed value is out of control and to suggest corrective measures.

The Combined Use of Control Charts and Correlation in a Quality Control Program

In the foregoing comments, the application of control charts and correlation have been discussed as separate aspects of a quality control program. It should not be construed from this presentation,



D

Mold Temperature		Mold Wash		Nozzle Condition	
Temperature	Index	Wash	Index	Condition	Index
A	2.4	A	2.2	A	3.1
B	4.3	B	6.1	B	2.4

Fig. 6—An Example of the Result of a Multiple Correlation Analysis. (A) Independent relation between iron content of slag and index of rejections on open-hearth steel. (B) Independent relation between casting temperature and index of rejections on open-hearth steel. (C) Independent relation between time in pits and index of rejections for surface quality on open-hearth steel. (D) Independent relation between nonquantitative variables and index of rejections on open-hearth steel.

however, that their application is disconnected. On the contrary it is only by a close co-ordination of the two, when dealing with operations such as we have in steel plants where isolation of factors causing trouble is extremely difficult, that a control program will develop its maximum efficiency. The control chart tells the state of control present in a variable but it does not tell whether the average level of

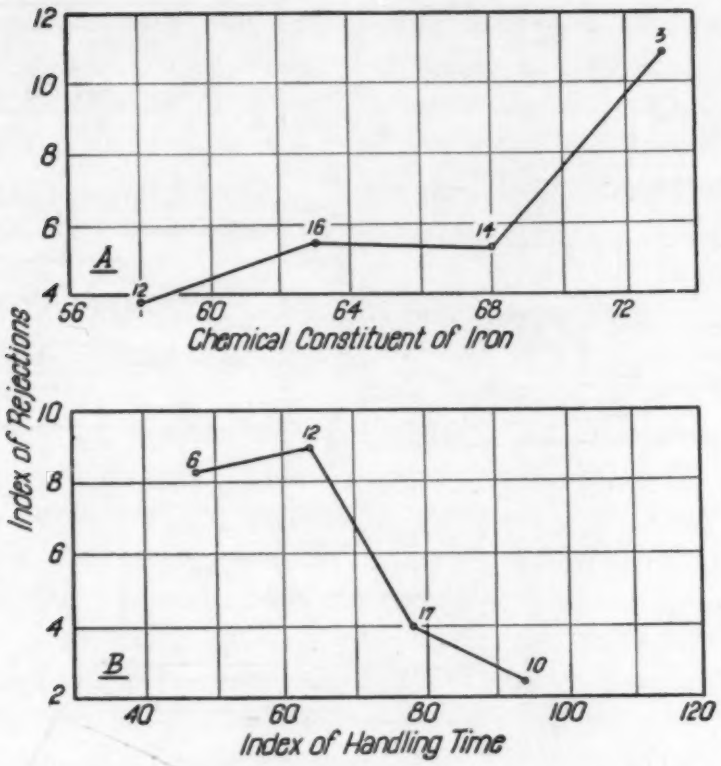


Fig. 7—Examples of Relations Developed by Correlation for the Purpose of Setting Up Control Procedure. (A) Relation between chemical constituent of iron and index of rejections. (B) Relation between index of handling time and index of rejections.

control and the variability observed is the condition desired. Evaluating the desired level of control is the function of correlation. It is therefore evident that the value of each method will be greatly enhanced by a co-ordination of their use. An example of the result obtained by this procedure is presented in the following problem:

It was desired, due to economic and production reasons, to change the method of processing a semi-finished steel product from one basic process to another. In setting up a standard practice for this new basic process the only practical engineering knowledge of factors related to the quality characteristics desired was that gained through experience on the old process. It became necessary, therefore, for the statistician and the engineer to resort to the three basic principles of scientific investigation, hypothesis, experiment and test of hypothesis. In this case, the hypothesis was based on the known facts in the old process and theoretical knowledge of the effect of variables on which no historical information was available. The experiment was the actual manufacturing of the product and the col-

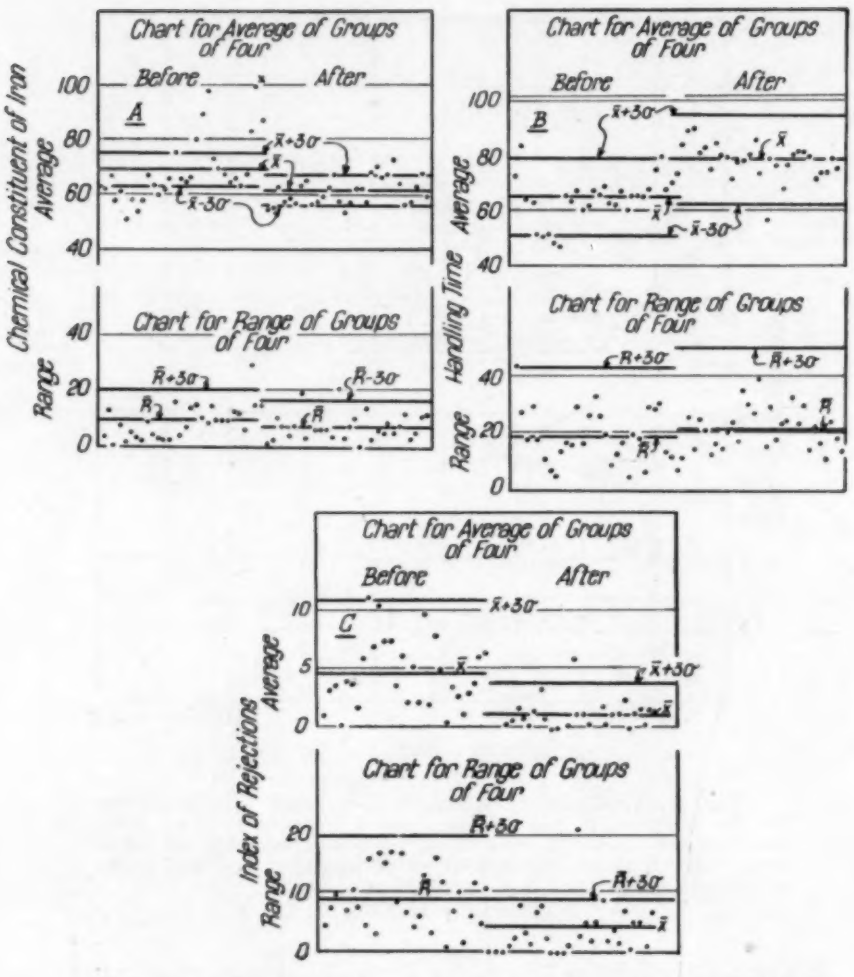


Fig. 8—Examples of the Result of the Application of Control Based on Correlation Analysis. (A) Control chart for average and range of chemical constituent of iron comparing conditions before and after control was installed. (B) Control chart for average and range of handling time before and after control was installed. (C) Control chart for index of rejections showing state of control before and after control was installed.

lection of all hypothetically related data. The test of the hypothesis was a statistical analysis by correlation and control charts. Variables known to affect quality in the former process were controlled at the level which had given the most desirable results and those on which no past information was available were left to fluctuate without restriction.

In problems of this kind it is not practical to wait until a large backlog of data has been accumulated before attempting an analysis of the various factors related to quality. For while thus marking time, product continues to be manufactured and if the quality characteristic is not controlled at a desirable level, continued production entails unnecessary loss of both material and money. Furthermore,

changes cannot be made promiscuously by the trial and error method because such practice is as likely to affect the quality of the product adversely as to improve it. In other words, making changes without foundation is unlikely to be more advantageous than pursuing a set course. Essentially the same problem is present here as in the example of multiple correlation presented previously. The probably related factors are numerous, the data are limited, and in addition it is necessary that results be known as soon as possible in order that rejects may be kept at a minimum and the best possible quality of product made.

There are various statistical procedures for attacking this type of problem such as the methods of analysis of variance, statistical principles of design of experiment and multiple correlation. In this case multiple correlation was the method employed with results as presented in the following brief discussion:

Twenty-four variables were evaluated with respect to their relation to the quality characteristic under consideration, indicated as "Index of Rejections". The relations developed with two of these factors, "Chemical Constituents of Iron" and "Index of Handling Time", are presented in Figs. 7-A and 7-B. It will be noted from this figure that the higher values of "Chemical Constituent of Iron" and the lower values of "Index of Handling Time" both tend to give a greater index of rejections. Based on the result of this analysis, specified limits were then set up for these and other variables which were found to be correlated to quality. Fig. 8 presents control charts showing the state of control of the quality related variables and the "Index of Rejection" before and after the correlation analysis was made. It is evident from these charts that the two variables under consideration definitely changed to conform to the findings in the correlation analysis and that this action was effective in materially reducing the index of rejections.

Conclusion

In the foregoing discussion an attempt has been made to enumerate the various phases of a complete quality control program as carried out in a large industry. In addition to the material presented there are various other statistical methods which are found applicable in specific cases. The applications of these are not routine and consequently not dealt with in this paper. These methods are avail-

able, however, in the many text books and the literature on the various uses of statistical technique in industry. In general, an attempt has been made to present in a nontechnical manner the broader aspects of quality control procedure by the use of statistical methods in one plant, to the end that the value of this procedure may be brought to the attention of those interested in this rapidly expanding science and its practical application.

ACKNOWLEDGMENTS

The author wishes to express his gratitude for the generous assistance of Messrs. J. D. Tyson, H. J. Hand, E. W. Earhart, R. McQuattie, J. L. Udvardy and other employees of the National Tube Company who have assisted in the preparation of this paper.

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DISCUSSION

Written Discussion: By D. K. Wright, Jr., assistant professor of mechanical engineering, Case School of Applied Science, Cleveland.

The author is to be commended for his paper which is evidently the result of considerable experience in the application of statistical methods to a manufacturing process which resists stubbornly the scientific approach and is still largely in the "art" stage. Two of the difficulties encountered in the making and processing of metals are, first, the problem of finding methods by which

phenomena may be measured; and, second, the complexity of the cause systems entering into the final product. For the first of these, ingenuity and experience combined with advances in the science of measurement must be depended on to furnish solutions; for the second, once a measurement is possible, statistical methods offer powerful aid.

The author has discussed four of the most widely useful of these methods in a manner to point out the application of each. In particular, he has given a demonstration of the efficiency of multiple correlation. As pointed out by the author, researchers prefer where possible to use the so-called Experimental Method which requires that data be gathered under conditions which hold constant all independent variables but the one whose effect is being investigated. This method affords a maximum of assurance for a minimum of testing. It is the method generally used by engineers, and the only one used by many of them. When such control over unwanted variables is difficult or costly, a statistical method such as Multiple Correlation or Analysis of Variance is usually preferable. The example presented in the paper brings this out in striking fashion.

It is the writer's opinion that the author, in his comparison of the frequency distribution with the control chart method, may have planted in the reader's mind the erroneous ideas that the two are alternative methods of attacking the same problems, and that the control chart method is always the better. This control chart is a device by which we attempt to judge on the evidence of the internal consistency of successive samples whether or not it is likely that the averages of those samples would fluctuate as they do without a change in the system of causes for variation. If it does appear likely, the control chart then gives a hazy picture of the parent population. The picture is hazy because we are dealing primarily with sample averages, and it is well known that the distributions of sample averages from skewed parent distributions are nearly normal. Thus parent distributions which are of dissimilar shape might give control charts which look alike.

The frequency distribution, on the other hand, makes no effort to judge the stability of the source of the data; but, rather, starting with the assumption that the data have come from a constant cause system, it gives us as good a picture of the shape of the parent population as it is possible to get from the data.

The two methods thus appear to be complementary rather than being competitors for the same job. The control chart will indicate to us whether the data are worth analyzing further. If a more accurate picture of the distribution is desired, a frequency distribution will supply it.

If some variable (usually time) is suspected of influencing the stability of a process, the control chart method is the one to use and the samples must be taken in such a manner (usually chronological order) as to permit testing of the question. This careful ordering of samples is the additional information which gives the control chart method its additional power. If all of the data were to be thrown into a hat (the frequency distribution method) and new samples drawn from it, the control chart method would lose its advantage.

In the analysis of experimental data, the control chart is again used for the purpose of evaluating the state of control existing during the experiment and therefore the worth of the data.

Beyond this, the control chart and the frequency distribution indicate about

the same situation. Figs. 3A and 4A both lead to the conclusion that the data for Condition A and those for Condition B have averages not far different and that Condition A is slightly more variable than Condition B. Neither the frequency distribution nor the control chart gives in this case the important information we are seeking; that is, how likely is it that these masses of data could be drawn by chance from the same distribution? For this problem, a statistical test of significance is indicated.

Again in Fig. 4B the purpose of the control chart is to indicate how good the control of the process was during the running of the test. As the author points out, there is some indication of increasing uniformity for both conditions toward the end of the test. Again the question of significance of any differences found remains unanswered.

Another application of statistics to quality control not touched by the author, and one which frequently leads to appreciable saving, is that of design of plans for sampling inspection. However, the writer realizes that the author was writing for limited space, and that a choice of subjects was necessary. He has given an interesting and valuable discussion of those he picked.

Written Discussion: By Earl W. Mahaney, metallurgist, Youngstown Sheet & Tube Co., Youngstown, Ohio.

Mr. Rogers is to be commended on the very splendid presentation of a subject not too readily digested by the average metallurgical practitioner or steel mill operator.

Industrial applications in the field of quality control, with the exception of a few companies, has been until recent years very limited. During the war, however, the use of statistical methods in solving many quality and production problems has been widespread. These methods have enabled the users to make substantial savings by quickly catching changes in quality levels of raw materials, excess material waste in lines of production, and in reducing inspection costs with adequate protection against the acceptance of inferior material still maintained.

However, it is our feeling that thus far the use of these techniques has been far too limited. Practically all of the literature available up to the present time treats problems dealing strictly with quality or with specifications. We have read much about such characteristics as percentage defective, deviations from aim-at chemistry, size tolerances, coating weights and thicknesses, and a host of others. It is true that each of these problems is of utmost importance and offers an unlimited field for the statistical evaluation of the observed daily fluctuations. For example, chemistry deviations may operate between certain limits which are dependent upon certain other variables such as material purity, handling times, human techniques, and perhaps many others. Some of these variables may be subject to control; others may not. Herein may lie a complicated problem involving the use of frequency distributions, control charts, and multiple correlation.

Is it not possible, then, that the tons produced per operating hour in a Bessemer plant or on an open-hearth furnace, or on a finishing mill, might be subject to fluctuations which are dependent upon a set of variables similar to those cited? It is quite possible that operating efficiencies could be markedly improved through studies of this type.

Another field of industrial application of statistical analyses has, so far as we know, not been touched—that of cost evaluation. Each of our companies has its own system of cost reporting, each with its own merits. Yet every department head has at times been at a loss to explain why a certain cost item is higher or lower than last month's figure. A statistical study of his previous figures coupled with other closely related items might show that the particular cost item in question was not significantly different from his past figures. Conversely, small fluctuations in monthly costs might be proven highly significant.

Mr. Rogers has not touched upon one statistical technique which we feel should claim an important role in any organized quality control program. When correlation studies have isolated troubles, and further methods such as control charts have helped to eliminate them, it follows that the final product has been improved. It then becomes practical to investigate the possibilities of reducing inspection costs or of eliminating some increment of the process. This can be done with the use of certain laws of probability in an acceptance sampling scheme. Our company has been able to reduce drastically the amount of conditioning done on semifinished bars and at the same time improve the over-all quality of the product by the use of a sample inspection and statistical grading system. Our system consists of the random selection of a relatively few samples from a heat of bars during rolling. A very careful inspection of these samples will send the heat either directly to the finishing mills, or to the conditioning yards, depending on the number and severity of defects found. We feel that the surface has barely been scratched in the use of quality control in all fields of all industry.

Written Discussion: By A. P. Bellinghausen, Kimberly-Clark Corp., Neenah, Wis.

Mr. Rogers' paper is a logically arranged and admirably presented argument for the use of statistical methods in control of quality. An extensive literature has been developed specifically for this purpose, but few of the articles make as good a case. The dice experiment shows very well the fact that faulty inferences can be made from a frequency distribution if, in effect, it is the sum of two or more distinct distributions. Technical men relying on a simple frequency polygon often have been found to overestimate the variability of product that is observed or may be expected from the best operation of a process. Several thoughts suggested by this paper may be of sufficient importance to warrant discussion.

When the author says that it is impossible "to determine definitely whether the different variables are in a state of control" from the frequency distribution, he places one in the position of not having available a tabulation of raw data arranged in the order of testing or production. If data are available in such form, several very simple tests of randomness may be used to indicate lack of constancy in the conditions generating the variable being measured. These tests are known collectively as the analysis of runs. Although not yet available in statistical texts, source material in several periodicals is cited in a Bell Telephone System Monograph (B-1319) by Dr. Shewhart. Some tests of homogeneity using run analysis are very sharp and some lack power but, in general, they are considered more useful than the type of analysis based on patterns in a control chart. From an industrial standpoint it is well not to draw a frequency polygon or calculate the parameter values of one unless the data have first met one of these tests for consistency.

For the second point in this discussion consider the following comment, relative to Fig. 2B: "—the last half representing the elimination of chance factors which were present in the first half." It is doubtful that the change giving this reduction in variability "took place", as the author suggests. More probably a change "was made" by altering quite materially some step in the process. The variability in product from a controlled process is a fixed characteristic of that process and can be estimated from samples of the product. Assume a process characterized by $\bar{x}$ and σ_T made up of two operations generating the variable $\bar{x}$ called operation A and operation B. Then the total variation can be explained in terms of variation due to A, and variation due to B:

$$\sigma_T^2 = \sigma_A^2 + \sigma_B^2$$

These values may be found and evaluated experimentally (under controlled conditions). If not in agreement with estimates made from data on the product, lack of control is indicated and assignable causes can be found and eliminated to eventually bring about agreement. But if data taken during production are found to be in control for total product and operations A and B, nothing will bring about a reduction in σ_T except a change in either A or B to an operation with lower variance.

If the data presented in Fig. 2B were actually used in the analysis then the statement that both sections "show control all the time" is open to question. Observe that of the first 31 averages plotted the first 15 seem to cluster much closer to the mean than those in the latter part of the first section and that the sample ranges seem to be somewhat lower in average range. This indicates inconsistency within Section 1. In fact, the first 15 sample ranges appear to fit the chart of the second section. In the second section, the sample averages do not exhibit control, the fourth point being outside of the new 3σ limits and, by estimation, 8 of the 30 points fall outside of the 2σ limits. It is expected that only about 5 sample means out of 100 should fall in the area of 2σ to 3σ. Thus inconsistency is exhibited in Section 2. There is little doubt that mill yield has improved and at the time indicated, but the reason was not the elimination of some "chance causes" and there is still work to be done.

For the third part of this discussion reference should be made to Figs. 3B and 4B. The author treats data which apparently were taken from mill operations under two different conditions, A and B, by plotting frequency polygons and control charts. It is almost impossible under operating conditions to get consistent experimental data and this imposes difficulties that often vitiate the experiments. It is usually a sounder approach to take less data under controlled conditions and analyzed by the usual significance tests than to use uncontrolled data and attempt a control chart analysis. The range charts of Fig. 4B show that data taken under both conditions were uncontrolled. The one maverick range point in condition B of the weight per foot data ($\bar{R}$-2.0) is contradictory to the statement that the results became increasingly uniform as the testing progressed.

In such investigation, one designs the experiment with condition A and condition B as the only two variables. Product made under condition A is expected to be homogeneous, else there is reason to suspect that other conditions which are identifiable have influenced the quality of product A. Product B must also be drawn from a controlled process. The experimentalist takes samples of

A and of B and performs his measurements. In analyzing the data he should first test each tabulation for consistency using all the information at his disposal, both engineering and statistical. The control chart and run theory are useful. If the data prove to be nonrandom perhaps they can be refined if the causes can really be found, or better still, the experiment should be repeated with the causes removed. When both sets are consistent, the parameters may be estimated from the data and the null hypotheses set up and tested. If accepted, the change in conditions cannot have affected the quality under consideration and if rejected the change in conditions is the only one that could have caused the shift in mean or change in variability. The point to be emphasized is that one cannot compare two distributions if, in fact, several exist during the course of the experiment. There is much danger in drawing far-reaching conclusions from experimental data taken under uncontrolled conditions. In this case lack of control could have been detected very early in the experiment by means of run analysis.

Mr. Rogers' paper will help to accomplish what he meant it to—to gain attention to and arouse interest in the use of statistical methods in problems of industrial quality. The contrast between analysis of data by plotting frequency polygons and by charting statistics of samples and detecting nonhomogeneity is strikingly presented. In the writer's experience many well trained technical men have been found to fall into the trap set by nonhomogeneous data and some have been found who had not acquired the concept of the distributive nature of quality. Technical people must learn to think with these concepts, especially in the metallurgical field, since almost every test specimen or every piece of material subjected to inspection is a sample drawn from some kind of a distribution. The ultimate goal of effort spent on control of quality is not to make each unit of product exactly like every other unit but to make products having the same distribution from lot to lot.

Oral Discussion

A. W. HARRIS:[2] I would like to ask Mr. Rogers if he has made an approximate estimation of the savings his statistical department has effected for his company? In order to sell the idea of statistical control methods to industry, this point must be considered.

Authors' Reply

In reply to Mr. Wright, it is true that the paper presents two methods: The frequency distribution and the control chart as alternative methods of attacking the same problem. It is also true that in a control program in a large industry, the control chart is by far the more valuable tool in following routine processing variables. As pointed out in the paper, there are things we must know about the performance of a variable before we are in a position to analyze it for causative behavior. For instance in Fig. 1A, would you be able to pick out any of the behavior characteristics of the data as shown in Fig. 2A, and in examining the two figures comparatively, which gives the clearer picture of what actually happens?

Is the cyclic occurrence of the data which is definitely evident in Fig. 2C to be observed in Fig. 1C? In industrial processing data we seldom have normal

[2]Research engineer, American Steel & Wire Co., Cleveland.

parent populations, and for this reason we are more interested in the control characteristics as time progresses than in the exact reconstruction of the data in the order of their frequency of occurrence.

With reference to experimental data, it is true that Mr. Wright's comment about Figs. 3A and 4A are correct, that we would hardly have made a decision one way or another from either analysis. Figs. 3A and 4A were not presented in this discussion, but they do not show much difference between a frequency distribution method and a control chart method, except they give you a picture of the state of control existing during the experiment and this is essential. In Figs. 3B and 4B, however, it was pointed out that we would have come to an erroneous conclusion by assuming that condition A was more variable than condition B, while from the control chart method, we would see that no conclusion should be made. It is pointed out that it keeps you from making conclusions which you should not make.

Replying to Mr. Bellinghausen, I perhaps should have cited the importance of collecting data in chronological order. In my own work data which do not have chronological identification lose much of their value.

I could state here that a change in the process "was made" after which a reduction in variability "took place". For the second part of this comment it can be stated that reductions in variability and shifts of an average occur as much from the elimination of chance causes as from an elimination of assignable causes and there is always work to be done whether a process shows exact statistical control or not.

It is correct that it is almost impossible to get controlled data under operating conditions but in our work many experiments are made while operations are proceeding. The point is, that with the control chart we are fully aware of what happens; whereas, from the frequency distribution, erroneous conclusions may be made. Analysis of runs as pointed out can be considered as further refinement of the control chart application and adds considerably to the technique of chronological analysis of data.

Mr. Harris's question is the basic approach to the acceptance of any new development. It is believed that our company is a pioneer in the application of statistical methods to the solution of steel plant problems. The Statistical Division of our Metallurgical Department has been operating for over 10 years. Management does not keep a tool this long if it is not economically advantageous.

DETECTION, CAUSES AND PREVENTION OF INJURY IN GROUND SURFACES

By L. P. Tarasov

Abstract

This paper brings together and reviews widely scattered information, published or otherwise, on injury in ground surfaces and supplements it by the results of the author's experimental work on this subject, mostly in connection with stresses in ground surfaces. With all this information in one place, it should be easier for a metallurgist to decide how to attack to best advantage problems involving possible or actual injury in a ground surface.

The methods of detecting cracks, stresses and burn in ground surfaces are discussed first. This is followed by a description of cracks and crack patterns that have been observed in practice, including both the cracks that occur spontaneously during or after grinding and the etch cracks that can be developed by suitable etching from stresses introduced into the surface during grinding.

With this as a background, the metallurgical factors are considered that have been repeatedly shown to cause hardened steels to be susceptible during grinding to trouble from cracking or from unduly high stressing. When retained austenite, untempered martensite or grain-boundary cementite are eliminated by improvement in the heat treatment, it is found that the susceptibility to these types of injury is greatly decreased; thus cracks and high stresses in a ground surface are not necessarily the result of poor grinding practice as is often erroneously assumed.

The numerous grinding factors are next taken up, and specific examples are presented regarding the influence of some of the more important ones upon possible injury to the ground surface. The extent to which surface stresses may be injurious is considered in the light of the subsequent use to which the ground part is to be put. Finally, methods are discussed by which undesirable stresses may be eliminated from ground surfaces after grinding.

A paper presented before the Twenty-seventh Annual Convention of the Society, held in Cleveland, February 4 to 8, 1946. The author, L. P. Tarasov, is associated with the Research Laboratories, Norton Company, Worcester, Mass. Manuscript received June 5, 1945.

Introduction

THE removal of metal by any kind of machining operation results in the plastic deformation of a surface layer and in the generation of heat, and the magnitude of the effects corresponds roughly to the rate of production. Economics dictates the employment of the fastest possible rate of working consistent with the minimum standard of satisfactory quality demanded of the piece in service. At high rates of working, it is impossible to cool the metal fast enough at the freshly formed surface to avoid high instantaneous surface temperatures. The most that can be expected is that the coolant keep the high temperature from penetrating much below the surface by removing the heat from it as soon as possible after the surface has reached the high temperature.

When the metal is hard, as in the case of hardened steels or of cemented carbides, the machining operation necessitates the use of abrasives, grinding being the method most frequently used. Just as with other types of machining, the grinding operation both stresses and heats the surface layer of the work, the magnitude of these effects depending on the severity of the process. However, it can be expected that the depth of the affected layer, which is generally related to the depth of cut, will be less for grinding than for an operation like turning.

A piece of hardened steel may be stressed highly enough by abusive grinding to cause cracks to form at those spots where the resultant stresses exceed the rupture strength of the metal. An improvement in the grinding equipment or procedures just sufficient to eliminate cracking would still leave the surface in a highly stressed condition. The important practical difference between these surface stresses and cracks is that the stresses are only potentially dangerous and their intensity can be diminished in various ways to render them harmless; cracks, on the other hand, constitute a permanent form of damage that cannot be healed except by removing the cracked layer of metal, and ordinarily this is not feasible.

These cracks, to which such terms as grinding cracks, grinding checks and heat checks have been applied in the past, may be invisible to the naked eye but several methods are available to bring them out more clearly. Cracks already present in the metal, especially those resulting from heat treatment, are sometimes overlooked until the surface is sufficiently brightened by grinding, to which

operation they are then mistakenly attributed. Such cracks, however, are outside the scope of this paper except for methods of detecting them and of differentiating between them and cracks related to the grinding operation.

It is now generally recognized that when cracks do appear in a ground steel surface, grinding is not necessarily the sole cause nor even the major cause. The condition of the steel, having to do both with its composition and its heat treatment, may in some cases be primarily responsible, with grinding severity being a minor factor by comparison. Cracks are objectionable regardless of their principal cause and to keep them from forming under production conditions requires that both the metallurgical and the grinding aspects be considered.

Stresses in a ground surface may or may not be harmful, depending on their nature and magnitude and on the use to which the part is to be put. If they are such as to cause cracking at some later time, they are obviously just as undesirable as actual cracks. Thus when a part is subjected in use to heavy alternating stresses, then surface stresses that decrease its endurance properties are objectionable, but the same part may be entirely satisfactory when less heavily loaded so that its service life is no longer affected by the stresses in the surface.

Burning of the ground surface is another possible consequence of grinding hardened steel too severely. As is well known, this may lead to a harmful softening of the surface, especially in the case of cutting tools. In special instances, burn may also be responsible for a decrease in endurance properties.

In spite of the practical importance of this whole subject, little has appeared in print and the information that has been published is widely scattered and hard to locate.* The few available articles are mainly descriptive in character and each covers only certain special portions of the general problem of injury in ground surfaces. Thus a metallurgist or grinding engineer interested in this subject has been forced to rely mostly on his own experience.

The purpose of the present paper is to bring together in one place the useful published information on injury in ground surfaces, the results of field experience, and the results of the author's experimental work. Topics discussed include methods of detecting, recog-

*A fairly complete list of papers (1-23) appears separately in the bibliography appended to this paper. The figures appearing in parentheses pertain to the corresponding figures in the bibliography.

nizing and avoiding the various types of injury, and their causes, nature and possible effects. Because of the extremely large number of variables present in grinding, the experimental work has been mainly exploratory in character. The results described apply to the particular experimental conditions under which they were obtained; however, they also indicate along what lines a metallurgist might well proceed in tracking down similar troubles arising under different grinding conditions.

It is particularly worth emphasizing that just as most pieces of heat treated steel are not injured by heat treatment, so grinding normally does not lead to any of the types of injury forming the subject of this paper.

Methods of Detection

Cracks in Steel—There are a number of ways in which cracks, whatever their origin, can be detected in the surface of a piece of hardened steel, even though the cracks are too fine to be visible to the naked eye or are covered over. When they are merely very fine, microscopic observation with either oblique or vertical illumination may reveal them, the former being advantageous in distinguishing between cracks and scratches.

Another method of detecting cracks in steel is by magnetizing the piece and causing extremely fine magnetic particles to collect at the cracks. This is the basis of the widely used Magnaflux method. Cracks that are covered over or are camouflaged by the scratch pattern can be revealed clearly in this manner; besides, even when the cracks can be found by visual inspection, the magnetic-powder method is superior in that it immediately focuses attention upon the defects. One possible drawback to this method is that nonmetallic inclusions near the surface may be mistaken for cracks because such inclusions are also discontinuities from a magnetic standpoint, causing the powder to collect in their immediate vicinity. Some experience with the magnetic-powder method is necessary if mistakes like this are to be avoided.

When the magnetic-powder method cannot be used for any reason, a number of etching procedures are available to reveal cracks that cannot be seen directly. The one that affects the surface the least is a short etch with 5 or 10 per cent nital. This etchant may discolor the surface but it does not remove any appreciable thickness

of metal and because of this it is not always capable of revealing very fine cracks or those that are covered over.

A much more effective etchant, which can be used to bring out any cracks that may be present, consists of 25 to 50 per cent nitric acid,[1] used cold. However, when the steel is immersed in this acid for only a few seconds, as recommended by Morrison (8), the cracks may still be fine enough to make the use of low-power magnification desirable for locating them. A more prolonged exposure to the acid will further widen the cracks and render them visible without magnification, but the surface will be eaten away more deeply. Just what conditions of etching are preferable depend on the circumstances.

Cracks can also be shown up very prominently by deep-etching the steel in either hydrochloric or sulphuric acid of suitable concentration. These two acids, as will be explained in detail below, introduce an entirely new factor into the picture. They not only enlarge cracks that are already in the steel, but they are also capable of developing *new* cracks from stresses present in the surface. These cracks we shall find it convenient to call *etch cracks*.

In order to avoid developing etch cracks when sulphuric or hydrochloric acid is used for detecting cracks already present in hardened steel, it is necessary, before any etching is done, to remove the stresses capable of causing etch cracks. This can be accomplished by subjecting the steel to a tempering treatment sufficient to stress-relieve it, as recommended in the METALS HANDBOOK (12). When it is desired to detect both cracks and stresses in the steel by means of these acids, one piece should be etched in the stress-relieved condition to show what cracks were originally present, and a duplicate piece should be etched without any stress-relieving treatment to give both the original cracks and the etch cracks. If only a single piece is available, a portion cut from it can serve as the duplicate.

In deep-etching for cracks, the usual macroetching technique is applicable, except that the time can be reduced to 5 or 10 minutes. Etchants commonly used are 50 per cent hydrochloric acid or 15 per cent sulphuric acid. Both are used hot, preferably between 160 and 180 degrees Fahr.; however, the sulphuric acid can also be used at room temperature, but its action will then be much slower and the crack width will be comparatively small.

The destructive nature of these deep-etching methods is clearly

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[1]This means a 25 to 50 per cent solution, by volume, of commercial concentrated nitric acid in water. Other etchants are similarly expressed in this paper.

evident, as is the nondestructive nature of the magnetic-powder method. The nital etch can be considered nondestructive under most circumstances while the nitric acid etch is more on the borderline. The superiority of the magnetic-powder method over visual or microscopic examination lies in its far greater sensitivity and this, together with its nondestructiveness, makes it the preferred method for detecting cracks in hardened steel surfaces, especially under production conditions.

Stresses in Steel—As we have seen, stresses in a hardened steel surface, such as those found after severe grinding, can be developed into etch cracks by immersing the steel in hydrochloric or sulphuric acid but not in nitric acid. This was first demonstrated by Brophy (2), who explained it on the basis of the added stresses introduced by the absorption of hydrogen. When hydrochloric or sulphuric acid attacks a steel surface, atomic hydrogen is given off and diffuses readily into the steel; with an oxidizing acid like nitric, however, the hydrogen is oxidized before it can get into the steel.

It is now generally accepted that atomic hydrogen diffused into steel recombines into molecular form when it gets into submicroscopic rifts or cavities inside the metal. Molecular hydrogen cannot diffuse through steel to any appreciable extent and it does not dissociate into the atomic form. Thus more and more molecular hydrogen is occluded inside the cavities as long as atomic hydrogen is available for diffusion into the steel. The resulting high pressures set up triaxial stresses that are responsible for the phenomenon of hydrogen embrittlement, the various aspects of which have been discussed at length by Zapffe and Sims (24), (25).

We can think of the etch cracks as being formed when the original surface stresses plus those resulting from hydrogen pressure exceed the original rupture strength of the steel; or we can picture the rupture strength as decreasing, as a result of hydrogen embrittlement, until the steel is no longer strong enough to withstand the surface stresses. The actual mechanism is not important for our purpose, and it is mainly as a matter of convenience that the etch cracks are considered as arising from the addition of hydrogen-induced stresses to the original surface stresses.

Experimental evidence indicates that only certain types of surface stresses can be developed into etch cracks. In the first place, a crack will form only from a tension stress, since a compression stress would tend to keep the metal closed up at that point. An example of

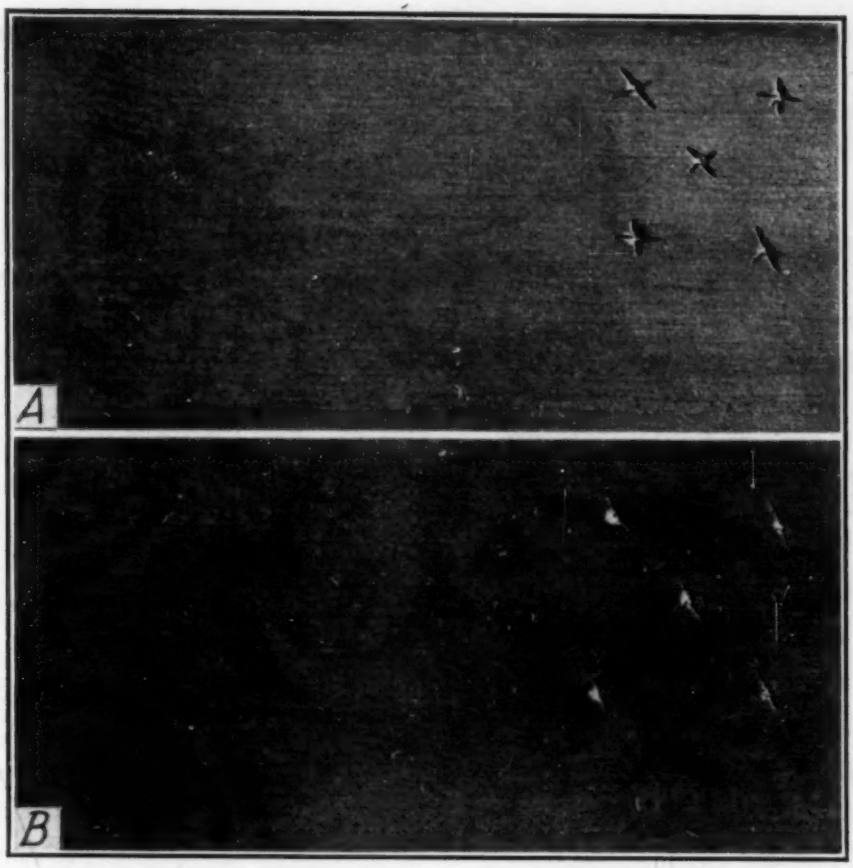

Fig. 1—Etch Cracks in Hardened Steel Developed from Stresses Introduced by Rockwell Indentations. Strips of S.A.E. 1095 steel were hardened to Rockwell C-66.5 and deep-etched for 15 minutes in acid at 170 degrees Fahr. (75 degrees Cent.). × 2. A—15 per cent sulphuric acid. B—50 per cent hydrochloric acid.

the effect of tension and compression stresses is furnished by the cracks often found radiating from a Rockwell indentation in a piece of fully hardened steel etched after the hardness was measured. As can be seen from Fig. 1, the indentations themselves are not cracked, since they are stressed in compression, but only the surrounding areas, which are stressed in tension, show cracking.

Another requirement appears to be that a stress concentration or gradient exists over a certain minimum distance. It is generally agreed that fully hardened tool steel is full of microscopic stresses, yet a small, regularly shaped piece of such steel can be drastically etched in hot hydrochloric or sulphuric acid without causing any etch cracks to appear. This is also illustrated in Fig. 1, where no etch cracks can be seen (other than those due to the Rockwell impressions) in S.A.E. 1095 steel hardened to Rockwell C-66.5 and not stress-relieved

after the brine quench. When etch cracks do appear in hardened steel unstressed by subsequent operations, it is because of large-scale stress inhomogeneities arising from nonuniform heat treatment.

When hardened steel is ground severely enough, the stresses developed in the surface are of the right type to cause etch cracks to appear upon suitable etching. As we have just seen, there are other sources of stress capable of leading to etch cracks, but these will be ignored hereafter and we shall be concerned only with etch cracks arising from stresses occurring in connection with the grinding operation.

In the discussion of the mechanism by which etch cracks are apparently formed, no mention was made of the possibility that selective etching at points of highest stress concentration might be the mechanism. This possibility, however, must be ruled out on the basis of some experiments in which atomic hydrogen was charged into pieces of hardened and ground tool steel without the use of any etchant. The steel served as the cathode at which atomic hydrogen was produced by electrolysis of a salt solution. The cracks that resulted from this cathodic electrolysis were indistinguishable from the etch cracks obtained in similar pieces by etching in cold sulphuric acid. The electrolytically treated surfaces were just barely discolored and there were no signs at all that any metal was dissolved from them, as is always the case in etching. These results show clearly that etch cracks form primarily because of the stress effects of the hydrogen charged into the steel and not because of selective solution of the metal in those regions that happen to have been highly stressed in grinding. If the latter effect does exist, it is decidedly a secondary one.

Comparison of Etchants for Developing Etch Cracks in Steel—Once an etch crack has formed, its appearance may be considerably altered as a result of continued solution by the etchant of the metal on both sides of the crack. As would be expected, the longer the etching goes on and the higher the etchant temperature is, the wider the crack will become. The nature of the etchant will also influence the crack width. These factors will likewise affect the rapidity with which the first cracks appear and the number of cracks that are developed in a given time.

These effects can be noted in Fig. 2, which shows eight severely ground pieces cut from a single ring of a hardened manganese oil-hardening tool steel and then etched for either 5 or 30 minutes in dif-

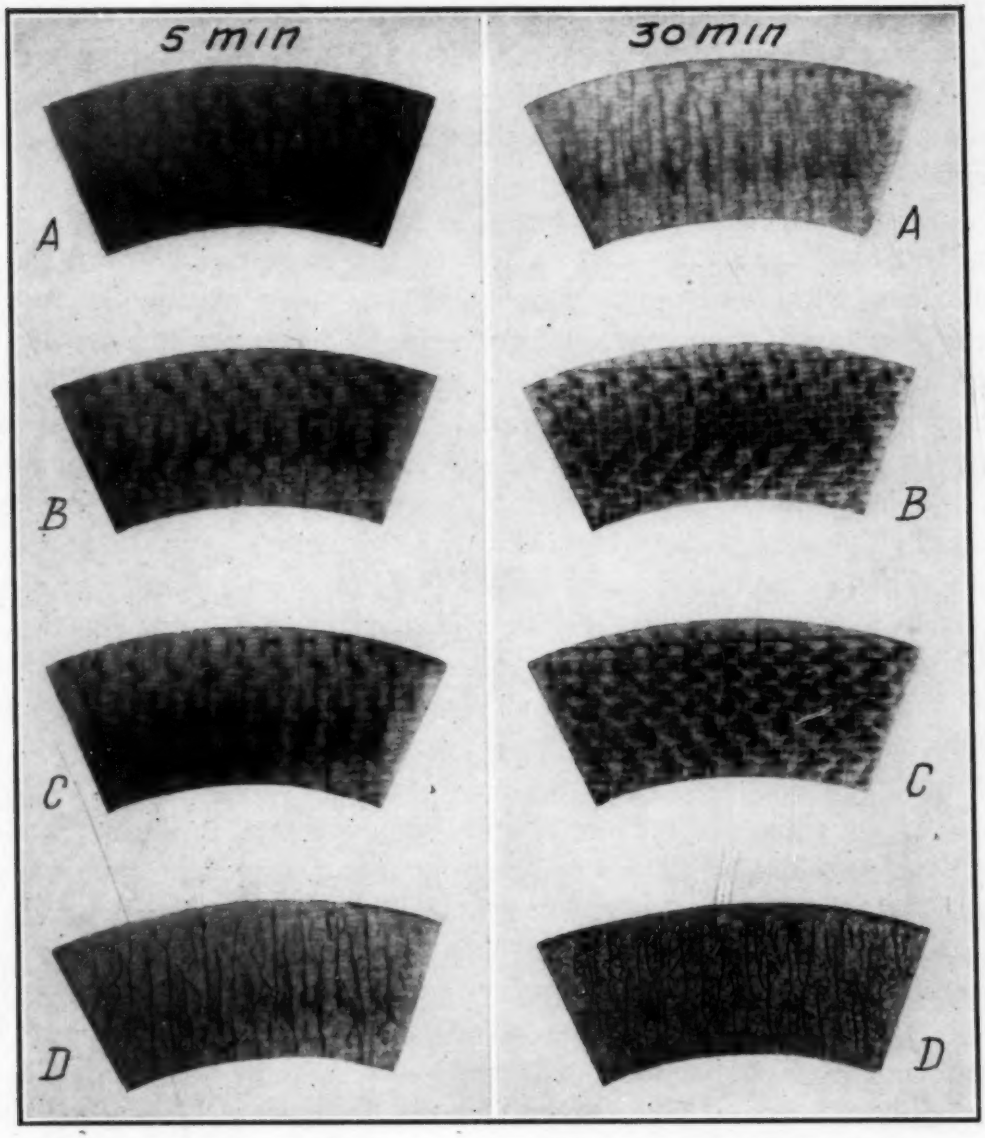

Fig. 2—Appearance of Etch Cracks Developed by Various Etchants in a Hardened Oil Hardening Tool Steel. The identically ground specimens were etched as indicated:

A—15 per cent sulphuric acid, cold.
B—38 per cent hydrochloric acid, 12 per cent sulphuric acid, balance water, hot.
C—50 per cent hydrochloric acid, hot.
D—15 per cent sulphuric acid, hot.

The etch cracks are the fine dark vertical lines, present in all the samples. The pattern of light and dark regions visible in most of the pieces is due to the chatter marks brought out by a light polish. × 1.

ferent acids. The acids were: Cold 15 per cent sulphuric acid; hot hydrochloric and sulphuric acids, consisting of three parts by volume of concentrated hydrochloric acid, one of concentrated sulphuric, and four of water; hot 50 per cent hydrochloric acid; and hot sulphuric acid, same concentration as the cold one. The cold acid was at about

75 degrees Fahr., while the hot acids were at about 160 degrees Fahr. The specimens were polished briefly on a wet metallographic lap with No. 600 Crystolon (silicon carbide) abrasive grain to remove most of the oxide film resulting from etching and thus to bring out the finer cracks more clearly.

All the specimens were found to be cracked after etching. The widest cracks, and the ones that stood out most clearly, were those obtained with hot sulphuric acid; at the other extreme were the cracks obtained with cold sulphuric acid. The cracks in some of the specimens did not reproduce too well in the photograph, although they were fairly easy to see directly. It should be kept in mind that cracks obtained under the same circumstances in a different steel could be wider or narrower than those shown, depending on how rapidly the steel dissolved in the various etchants.

That hot sulphuric acid makes the etch cracks stand out more clearly than does hot hydrochloric acid can also be seen in Fig. 1. This has been found to be true in steels as diverse as plain carbon, oil hardening, and high speed tool steels. The hot sulphuric acid not only results in wider etch cracks but it also causes them to appear in a shorter time.

If we make the very reasonable assumption that the surface is stressed less by gentle than by severe grinding, experience has shown that hot acid can develop etch cracks in pieces of hardened steel that are only moderately stressed by grinding and in which etch cracks cannot be made to form by immersion in cold acid. Increasing the etching time has been found to have the same effect as increasing the temperature of the etchant, but to a less pronounced degree. Similarly, the amount of etch cracking under constant etching conditions has been found to increase when the surface stresses increase. These effects were first observed by Brophy (2) and Morrison (8) and have been confirmed on numerous occasions by the results of the present work.

Of the etchants listed above, cold sulphuric acid is the most convenient to use in extended laboratory investigations, especially if many specimens are to be studied at a time, provided that the surface stresses are high enough to develop into cracks in a reasonable time. Another advantage of the cold acid is that it can be used over many times, under a condition to be described, while the hot acids dissolve so much steel that they have to be discarded after comparatively little use. Hot acid is harder to handle than cold and requires

more equipment, its temperature has to be controlled within fairly narrow limits, and the specimens should be preheated if the results are to be reproducible. The advantages of the hot acids are that they bring out stresses too low to be developed into cracks by the cold acid and that the cracks are wider and thus more easily visible.

The effect of concentration of hot sulphuric acid was studied by Brophy (2), who found that 15 per cent acid caused more etch cracking, other things being equal, than either the 5 per cent or 30 per cent acid. For cold sulphuric acid, the writer observed a similar but not very pronounced effect. There was a considerable range of concentration in which the maximum cracking was obtained and as the 15 per cent acid was somewhere in the middle of this range, it was chosen for all subsequent work.

In order to avoid replacing this cold sulphuric acid any oftener than necessary, a study was made of how many identically ground pieces could be etched successively for 15 minutes in a 100-ml portion of the etchant without any marked change in the amount of cracking. Surprisingly enough, the amount of cracking, measured in terms of the total length of cracks, rose rapidly for the first few pieces and then became approximately constant, as shown by the circles in Fig. 3. The abscissa is expressed in terms of the etchant age, which is the number of minutes that the 100-ml portion of etchant had previously been in use on similar pieces. Wherever more than 15 minutes elapsed between successive points, a dummy specimen was used to age the etchant correspondingly. The width of the specimen was taken as the arbitrary unit of measurement for the total crack length.

Since it appeared reasonable to associate the increasing amount of cracking with the increasing ferrous-ion content of the etchant, 15 grams of ferrous sulphate ($FeSO_4 \cdot 7H_2O$) were added to one liter of fresh 15 per cent sulphuric acid, this being equivalent to the ferrous-ion concentration at the knee of the curve just described, as found by chemical analysis. When the test was repeated with this new etchant, the results fell along a straight line passing through the crosses of Fig. 3. This line is almost horizontal and for all practical purposes it coincides with the nearly horizontal portion of the other curve. As drawn, the straight lines slope slightly downward with increasing etchant age, but it is uncertain whether this is a real effect or one due to experimental error.

Thus we see that the crack-inducing power of cold sulphuric acid rises to approximately double its initial value while the ferrous-

ion concentration is being built up during use to a certain critical value, and beyond this point the crack-inducing power remains practically constant. The addition of the critical concentration of ferrous ions to the unused etchant ages it artificially so that the crack-inducing power is at its maximum from the very start. It is for this reason that it is advisable, if cold sulphuric acid is to be used, to stabilize it with the proper amount of ferrous sulphate.

Chemical analysis at various stages of use of cold sulphuric acid not treated with ferrous sulphate showed that the ferrous-ion concentration increased linearly with etchant age so that the rate of solu-

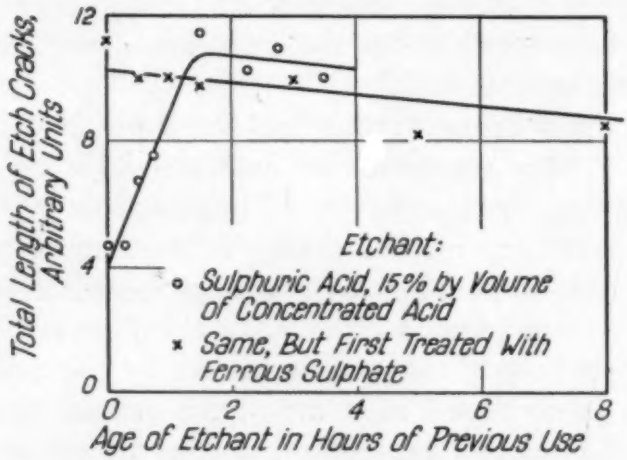

Fig. 3—Effect of Ferrous-Ion Concentration in Cold Sulphuric Acid on Development of Etch Cracks. Specimens like those in Fig. 2 were etched successively for 15 minutes in the same 100-ml portion of etchant.

tion of the steel was constant. Thus it seems very likely that the function of the ferrous ions is to increase the diffusion rate of atomic hydrogen through the surface of the steel. An increased diffusion rate would lead to greater stressing in a given time and consequently to an increased amount of cracking. An analogous effect may be that reported by Baukloh and Zimmermann (26), who found that elements like arsenic greatly increased hydrogen absorption from a certain electrolytic bath.

Other Methods of Detecting Stresses in Steel—A totally different method of studying surface stresses involves the measurement of dimensional or shape changes in a suitably prepared specimen when the stresses are permitted to redistribute themselves by the careful removal of some particular portion of the specimen. One method of this kind has been recently devised by Almen (18) for measuring various kinds of surface stresses in flat strips, including those stresses

introduced during grinding. A thin strip of steel, which has been properly stress-relieved throughout, will bend upwards at the ends when the upper surface has been stressed in tension by some operation. By carefully honing off successive extremely thin layers from the top surface and measuring the corresponding changes of curvature, it is possible to calculate the distribution of stresses originally present in the stressed surface. In one such experiment, Almen found that a very thin surface layer of severely ground annealed spring steel was highly stressed in tension.

Another experimental procedure of the same general nature has been used by Greaves, Kirstowsky and Lipson (23), who ground the outer surface of a thin steel ring and measured the change in diameter after cutting it parallel to its axis. In this manner, they also found that under their grinding conditions the ground surface was stressed in tension.

Although experimental proof is lacking, it is likely that these tension stresses are associated with the thermal contraction of small surface regions which follows the thermal expansion during momentary contact with the grinding wheel. The restraint placed upon thermal expansion by the adjacent cold metal results in compression stresses that are relieved by plastic flow perpendicular to the surface. This upsetting action takes place with increasing ease as the temperature goes up since higher temperature causes a decrease in the yield strength. As soon as cooling starts, tension stresses begin to be built up as a result of thermal contraction but now they can no longer be almost completely relieved by plastic deformation, as was the case previously, the reason being that the yield strength becomes high as the temperature drops. Thus the surface stresses, which are in tension, may be of any magnitude up to the room-temperature yield strength. This picture of the origin of tension stresses in a ground surface is essentially the same, aside from the dimensions involved, as that for stresses set up in a disk heated in the center to a high temperature (27), (28).

The question occasionally arises as to whether the magnetic-powder method could be applied to detect stresses introduced by grinding. The advantage of such a method would lie, of course, in its nondestructive nature, in contrast to the destructiveness of the etch method and to the necessity of using special test pieces if change in shape is the criterion. Unfortunately, the magnetic-powder method is incapable of detecting stresses or stress concentrations in the

surface of a piece of hardened steel. To cite an example, a small piece of hardened steel was ground just severely enough for a single surface crack to become visible while the piece was being polished for metallographic purposes. Even when this crack was covered over by flowed metal as a result of the polishing operation, its presence could be immediately detected by the magnetic-powder method. No other cracks could be located, either by the magnetic-powder method, or by microscopic examination after all flowed metal had been removed by repeated polishing and etching with nital. However, numerous etch cracks were developed as soon as the specimen was etched briefly in cold sulphuric acid. Only when the stresses had developed into actual cracks was it possible to locate them by the magnetic-powder method.

Even though the etch method for detecting surface stresses is destructive because of the dissolving effect of the acid, it can be very useful from a practical point of view. In many cases it is worth while to destroy a few pieces out of a large number to determine heat treating and grinding conditions that will not leave the work surface stressed too severely.

Burn in Steel—A sign of severe grinding is the appearance of burn on the surface as a result of high grinding temperature. This burn is nothing other than a very thin film of oxide, exactly the same as is responsible for temper color. Associated with burn in hardened steel is a softened layer whose depth and hardness depend on the circumstances. The relative severity of grinding can often be told from the color of the burn, a barely noticeable light brown burn changing gradually to a deep blue with increasing severity. However, severe grinding is not necessarily revealed by visible signs of burning since the metal being ground may be resistant to oxidation or else the burn may be cleaned off by the wheel, either practically as soon as it is formed or during the final finishing passes.

Even when visible signs of burn have been removed from the surface of a piece of hardened steel, the effects may still persist in a thin surface layer, and these can be brought out by etching briefly in 5 or 10 per cent nital or in nitric acid. These etchants will darken the surface in those spots which were burned for the same reason that a metallographic specimen of hardened steel, upon being etched in nital, will be darker the higher the temperature at which it was tempered. Hydrochloric and sulphuric acids can also be used to reveal burn.

The presence of burn and the depth to which it extends can be

easily detected in a metallographic section taken perpendicular to the burned surface after the section has been etched with nital.

Recommended Procedure for Detection of Injury in Ground Surfaces of Steel Parts—When a ground surface of hardened steel is to be exhaustively examined for signs of possible injury, the first step is to look for cracks. Assuming that none are visible to the unaided eye, the magnetic-powder method is the most satisfactory one for detecting them. The next step is to test for burn by subjecting the steel to the nital etch. This etch can also be employed in looking for cracks if the magnetic-powder method is not available for the purpose; however, if nital does not reveal any cracks, it becomes necessary to try nitric acid to make sure that cracks are really non-existent.

In case no cracks can be found, then the possibility of stresses in the ground surface should be investigated. By suitably varying the etching conditions, progressively lower stresses can be developed into etch cracks. Thus for some steels, such as plain carbon or oil-hardening tool steels, a 15-minute etch in cold 15 per cent sulphuric acid will bring out the highest stresses; and prolonging the etching action will cause progressively lower stresses to transform into etch cracks. Etching for 10 minutes in hot sulphuric acid will generally reveal the lowest stresses that can be made to appear as etch cracks. It is thus possible to get a rough idea of how severely the surface is stressed in tension by noting under what etching conditions the first etch cracks appear.

Cracks and Stresses in Other Hard Metals—Thus far, everything that has been said about methods has referred only to hardened steel. By comparison, little is known about the detection of cracks that may appear after the improper grinding of other hard metals, like cemented carbides, tool materials of the Stellite type, and magnetic materials of the Alnico type. As for surface stresses introduced into these hard metals during grinding, no methods have been developed for detecting them so that nothing at all is known about such stresses.

In cemented carbides, cracks are usually very difficult to see directly unless they are so serious that they are accompanied by spalling of the corners. Microscopic examination may reveal the cracks but in general they are likely to be smeared over. Some modification of the old kerosene and whiting method is frequently useful. A procedure that works fairly well involves soaking the piece of cemented

carbide for half an hour in hot spindle oil diluted with an equal portion of kerosene; this permits the oil to penetrate into the cracks. After the piece has been taken out of the oil and the surface has been wiped clean, it is dipped in a suspension of whiting ($CaCO_3$) in alcohol. This evaporates readily and leaves a fine deposit of white powder which can be freely brushed off except where the particles have adhered to the oil seeping out of the cracks. Variations of this method include the substitution of substances like levigated alumina or rouge for the whiting.

A more sensitive version of the same general method, known as the Zyglo method, is based on the detection by its fluorescence in ultraviolet light of oil seeping from the cracks (17). Here again it is found necessary to let the cemented carbide soak in hot oil; otherwise, the oil will not penetrate into the extremely fine cracks that may be present. A special emulsified penetrant can be used in place of the oil.

Cracks in cemented carbides can also be detected by etching for 3 minutes in a hot solution containing one part of concentrated nitric acid and two parts of concentrated hydrofluoric acid (19). This is somewhat more sensitive than the fluorescent-oil method and is capable of detecting cracks that can be located in no other way. However, this is essentially a laboratory method on account of the precautions that have to be observed when handling hydrofluoric acid.

A method that has seen some use in laboratory research involves the cross-bending strength of an accurately ground rectangular bar. The presence of cracks in the surface lowers the cross-bending strength and it may be that the presence of high tensile stresses in the surface does the same, but there is no proof of the latter as yet.

The cemented carbides are only very slightly magnetic even though they are bonded with cobalt, a magnetic element. Thus the magnetic-powder method is not applicable to the detection of cracks except in rare cases, and even then the fluorescent-oil method is far more sensitive.

As regards hard metals other than cemented carbides, the methods for detecting cracks are those already described. Alnico, which is magnetic, can best be examined by the magnetic-powder method; on the other hand, most grades of Stellite and similar cast tool materials, which are nonmagnetic, call for the use of the fluorescent-oil method.

Some Characteristics of Cracks and Stresses in Ground Surfaces

Cracks Present Previous to Grinding—If all the material that is to be ground were to be examined prior to grinding for the presence of possible cracks, such as those caused by improper heat treatment, then there would never be any question as to when cracks discovered after grinding were formed. Since such an examination is not general practice, it unfortunately happens sometimes that cracks already present but completely hidden from view, as by scale on the surface, become clearly visible after grinding and are then blamed upon this operation. For this reason, it is particularly desirable to be able to distinguish by their appearance cracks formed before from those formed during or after grinding.

Cracks present prior to grinding are generally irregular in shape and are likely to be related to changes in section, re-entrant or sharp angles, corners and holes; moreover, they are entirely independent of the direction of the scratches left by the abrasive particles in the grinding wheel, commonly called grain marks. They also tend to be much deeper than cracks occurring after grinding, which are quite shallow, and to be more widely separated. Further information on the appearance of cracks not related to grinding can be found in the literature (11), (12), (16), (29).

Cracks Developed During or After Grinding—The distinguishing features of these cracks are the shallow depth and the pattern, which is related in one way or another to the grain marks. Several pattern types are to be found in practice, depending on the circumstances under which the cracks were formed. In this discussion, the assumption has to be made that the grain marks are unidirectional for if they happen to be multidirectional, as occurs with some methods of grinding, it is obviously difficult to describe crack pattern in terms of scratch direction.

The simplest kind of pattern is one in which the cracks are roughly perpendicular to the grain marks. Cracks found in flat ground surfaces are most commonly of this nature. Typical examples are illustrated in Figs. 4 and 5. As would be expected, the direction of a crack is influenced to some extent by its proximity to the edge of the piece, the crack tending to turn perpendicular to the edge. The same change in direction is often observed as a crack approaches a hole.

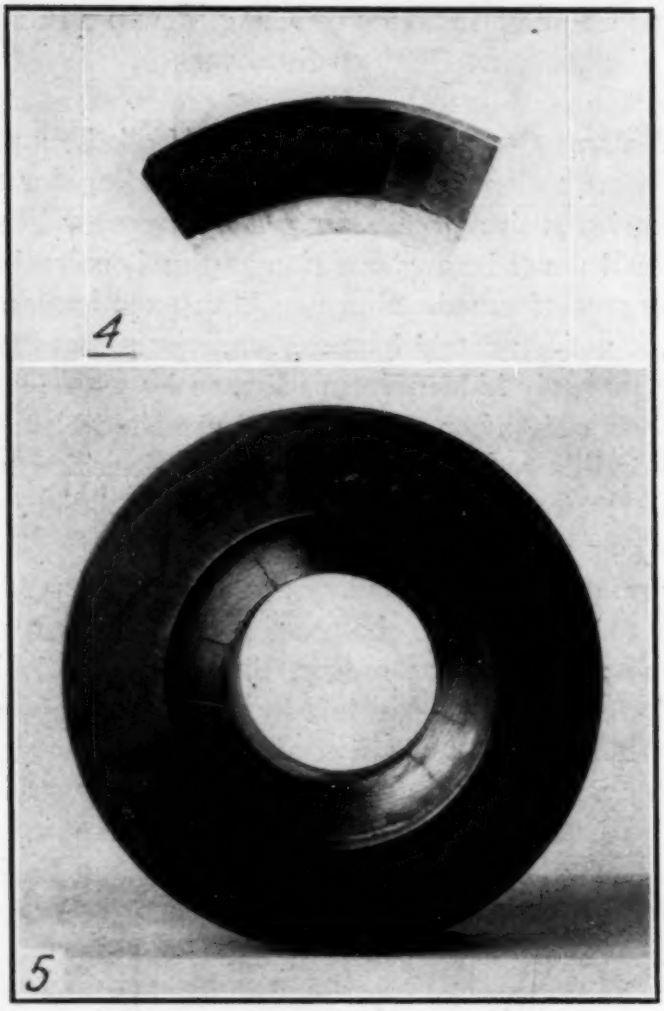

Fig. 4—Cracks Caused by Improper Grinding of Flange of Carburized and Hardened Steel Bushing. The radial cracks are perpendicular to the circular grinding direction. The light portion below the dark cracked area was ground down to the level of the flange in preparing the piece for photographing. × 1.

Fig. 5—Cracks Caused by Improper Grinding of Hardened Ring of Carbon Steel and Made More Easily Visible by Magnetic Powder Method. × 1.

Increased severity of grinding can lead to a branching of some of the cracks, such as can be seen in Fig. 4. When the branching is frequent enough, the cracks begin to form an irregular network in which the most pronounced cracks remain more or less perpendicular to the grain marks. With further increase in grinding severity, the network is likely to become a regular one having the cracks both perpendicular and parallel to the grain marks. This can be considered to be a second type of crack pattern in ground surfaces, but no sharp dividing line can be drawn between it and the first type

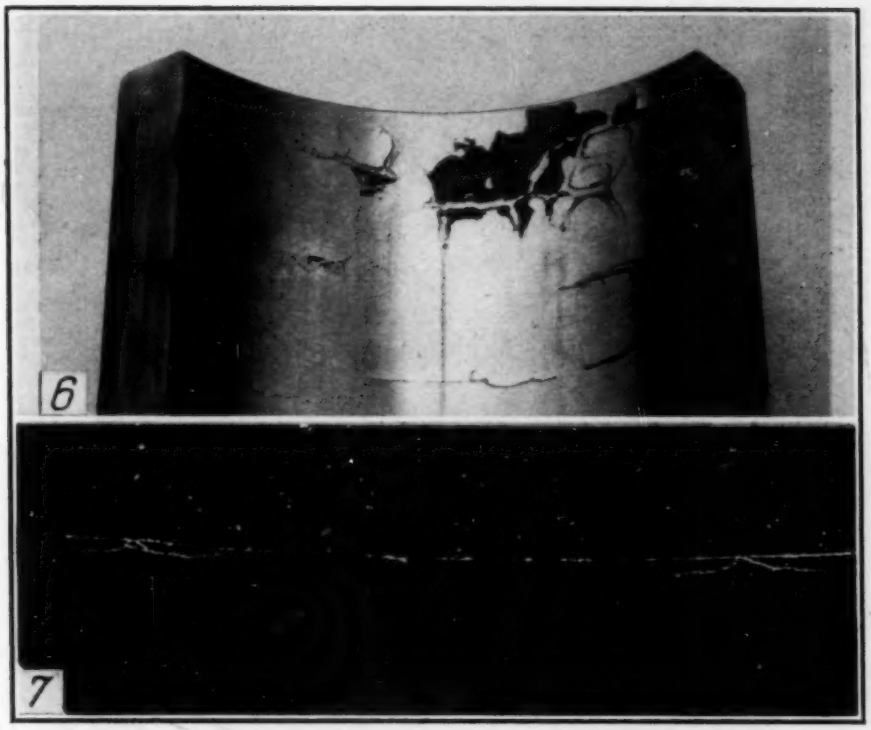

Fig. 6—Spalling of Internally Ground Surface of Carburized and Hardened Hollow Cylinder. × ⅔.

Fig. 7—Spalling Cracks of Fig. 6 in Section Perpendicular to the Cracked Surface and Located in Region Just Beginning to Crack Visibly. The cracks contact the slightly curved horizontal line separating the steel below it from the Bakelite mount above it. Illumination and exposure conditions adjusted to bring out the shape of the cracks most clearly. × 10.

of crack pattern since one type gradually merges into the other.

The third type of crack, occurring mostly in cylindrical surfaces, is essentially parallel to the grain marks, but considerable deviation and branching may be found in some cases.

The fourth and last type is the one that leads to spalling, an example being shown in Fig. 6. The important difference between these cracks and those previously described is that these become parallel to the ground surface at only a short distance from it as shown in cross section in Fig. 7; the cracks of the other three types, on the other hand, do not deviate much from their original direction perpendicular to the surface.

Cracks in ground gears are sometimes of the spalling variety, as illustrated in Fig. 8. A condition is shown in Fig. 9 in which the spalling cracks are not as well developed and in which some appear to belong to the first type discussed, i. e., those perpendicular to the grain marks. Both cracked gear teeth came from the same large gear and the difference in crack pattern must be ascribed to a varia-

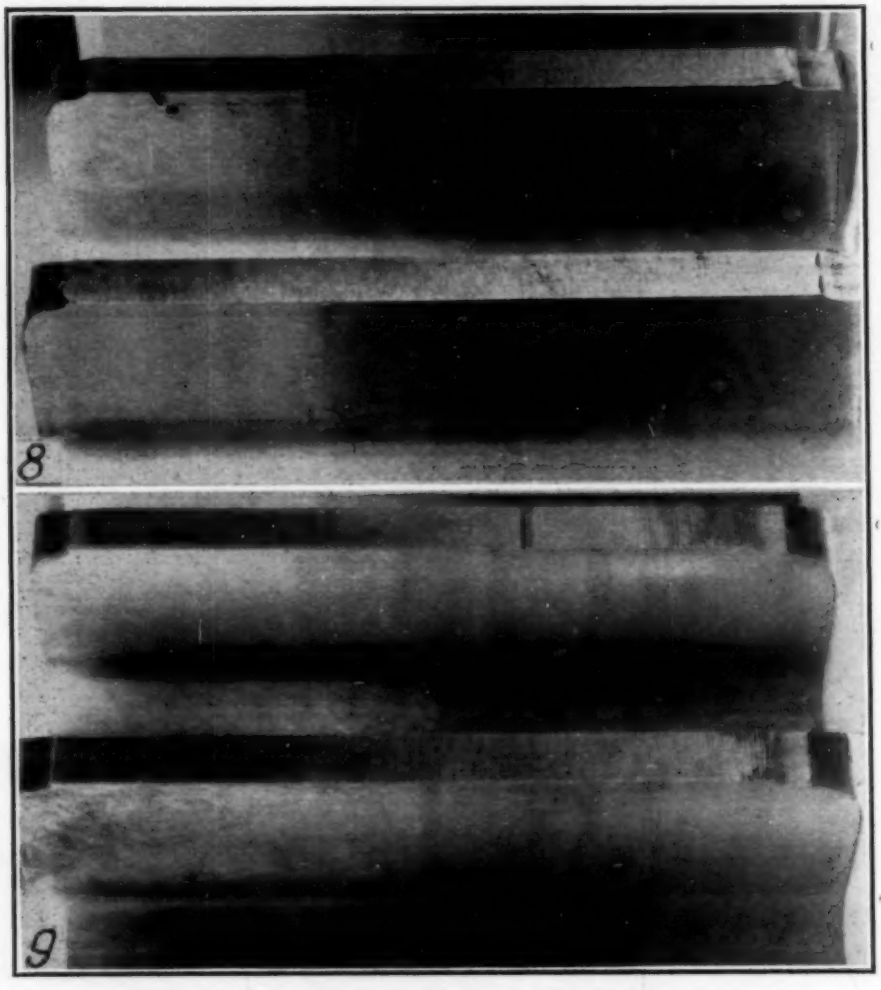

Fig. 8—Spalling Cracks at the Root of a Large Unfinished Gear Tooth, Caused by Improper Grinding. × ½.
Fig. 9—Cracks in Another Tooth of Gear Shown in Fig. 8, Also the Result of Improper Grinding. × ½.

tion in grinding conditions, the tendency to form spalling cracks increasing with severity of grinding. Some evidence of spalling cracks can also be noted in the lower right-hand portion of the specimen in Fig. 4. In general, it appears that spalling cracks are associated with severe grinding under conditions of broad contact between wheel and work.

Spalling cracks are considered to be due to the same forces that cause a thin strip of metal to curve after it has been ground severely. Because of thermal expansion and contraction of the surface layer, it is thought that there is a plane of weakness more or less parallel to the surface, which is unable to withstand a stress redistribution initiated by the formation of cracks starting at the surface.

Cracks developed during or after grinding extend to a depth which generally is no greater than 0.010 or 0.020 inch, although depths several times as great have been observed in special instances. When a piece of hardened steel is cracked all the way through as a result of improper grinding, it is highly probable that the cracks directly attributable to grinding were themselves shallow but that the steel was already too brittle and highly stressed to withstand the presence of any surface cracks.

Etch Cracks—The same patterns are exhibited by etch cracks in a hardened and ground steel surface as by cracks that occur in the absence of etching. This is to be expected since the stresses determining the pattern are the same in the two cases, only in one case the stresses are high enough to result in spontaneous cracking, while in the other, additional stresses caused by suitable etching are needed before cracking can occur. Typical etch cracks are shown in Figs. 2 and 10, where the horizontal direction of grinding is still visible after the etching treatment. Pieces such as the ones in these two illustrations were surface-ground by traversing them in the longitudinal direction.

Etch-crack patterns can be of various degrees of complexity, depending not only on the grinding conditions but also on the etching conditions. Some typical patterns resulting from progressively more severe grinding conditions are shown in Fig. 10. Branching is evident in specimen C and the beginnings of a network in specimen D. In this latter case, the network is not predominant enough to obscure the presence of the vertical cracks. When a network of such cracks appears after etching, it is found that etching a similar piece for a much shorter time results in a few cracks that are perpendicular to the grinding direction, and that the network develops only gradually with increased etching time. Since cracks form perpendicular to tensile stresses, these stresses in directions more or less parallel to the grinding direction are higher than those perpendicular to it.

In connection with stress levels, it is worth noting that a short surface crack appearing spontaneously after grinding will generally be considerably lengthened by suitable etching, the extensions of the original crack being regular etch cracks. This happens because the region immediately beyond the original crack has also been stressed by grinding, but not quite enough to form a crack without further stressing by etching.

Etch cracks are generally likely to extend no more than about

0.010 or 0.015 inch below the surface, but in exceptional cases they have been found to be several times as deep. This applies both to the narrow cracks typical of cold sulphuric acid and to the very wide ones caused by etching in hot hydrochloric or sulphuric acid. The latter appear to be much deeper because it is easy to see down to the bottom, but the depth is essentially the same. This was shown for specimens like those in Fig. 2, both by gently grinding away the cracked surface until no further traces of cracks could be found following a nital etch to remove any flowed metal, and by a direct metallographic examination of a suitable cross section of the cracked piece.

Occasionally etch cracks may be found that extend to the edge of the ground piece and appear to be fairly deep; however, sectioning shows that the crack is everywhere of normal depth but has extended itself over the side. Since the extension usually remains in the plane perpendicular to the ground surface, it is easy to mistake this extension for a cross sectional view of the original crack.

When etch cracks are developed in a flat ground surface by treating with cold sulphuric acid, and the surface, now oxidized dark by the etchant, is polished briefly on a metallographic lap as already described, it is found that each crack is located in the middle of a polished ridge, which appears bright against the rest of the surface. The height of a ridge above the rest of the surface is in the neighborhood of only 0.00005 inch. These bright ridges are very useful in showing where the etch cracks are located for very often the cracks themselves are visible only under magnification.

The ridges have been observed to form just prior to the development of a visible crack. When a suitable surface is etched long enough to bring out a few etch cracks and is then polished, it is sometimes found that cracks cannot be detected in some of the faintest ridges even under high magnification. The expected cracks appear only if the surface is etched a short while longer. Cracks formed prior to etching, on the other hand, do not have ridges associated with them. If a surface already containing cracks is etched and polished as described, the cracks originally present will be conspicuous by the absence of any bright ridges, while any etch cracks that have developed will be surrounded by ridges as usual. All this indicates that the ridges are somehow dependent for their formation upon stresses perpendicular to the surface and set up by the pressure of the hydrogen absorbed by the steel.

Similar ridges can just barely be detected under favorable cir-

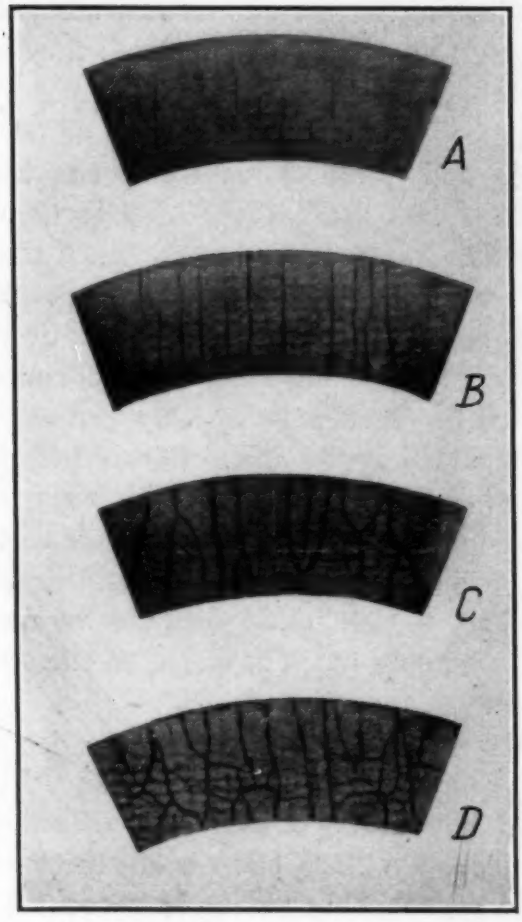

Fig. 10—Effect of Increasing Severity of Wheel Action Upon Etch-Crack Pattern in Hardened Oil-Hardening Tool Steel. Specimens traverse-ground in succession without redressing the wheel, causing the wheel to become progressively duller in action. Grinding was done on a reciprocating-type surface grinder so that the grinding marks are straight horizontal lines. Etch cracks developed in cold sulphuric acid were made more easily visible by magnetic-powder method. × 1.

A—Specimen No. 2 in series.
B—Specimen No. 6 in series.
C—Specimen No. 17 in series.
D—Specimen No. 20 in series.

cumstances when a hot etchant is used. In this case the surface is so rough and deeply oxidized that the ridges become only a trifle brighter than the rest of the surface; besides, the height of the ridges relative to the rest of the surface is probably decreased as a result of the comparatively large amount of steel dissolved during etching.

Neither kind of crack is necessarily associated with the chatter marks often observed on severely ground surfaces. Although cases

have been noted in which such cracks run along the peaks of the chatter marks, there are other cases in which they are located in the valleys, and still others in which they are completely independent of the chatter marks.

Stress Depth—It is naturally of interest to know the depth of the grinding stresses that are capable of being developed into etch cracks. The effective stress depth for hardened tool steel specimens like those in Fig. 10 was determined by grinding a group of them severely enough to result in etch cracks when treated in cold sulphuric acid. Before etching, however, various small amounts of material were removed from the specimens by very gentle grinding and only then were they immersed in the acid. It was found that etch cracks could not form in those specimens from which more than 0.0005 inch had been removed. This can be taken as the effective stress depth, at least for the conditions of this experiment. Since the etch cracks themselves were 0.010 to 0.015 inch deep for the most part, it is evident that they extend many times deeper than the stresses responsible for their formation.

There is evidence that an effective stress depth of 0.0005 inch occurs when entirely different metals are ground. Almen (18) found the region stressed in grinding a long strip of annealed spring steel to be about this deep. Wulff's electron-diffraction investigation of 18-8 stainless steel showed that the layer deformed by grinding extended to a depth of almost 0.0005 inch (15). All three cases here referred to involved dry traverse grinding on a reciprocating type of surface grinder. Considerably more data on effective stress depth under different grinding conditions are needed before the significance of the 0.0005-inch depth can be properly evaluated.

In the work on 18-8 stainless steel just referred to, a variety of metal-removing operations, including wet and dry grinding, were studied by Wulff. In this type of stainless steel, deformation transforms austenite to ferrite, at least partially, below about 390 degrees Fahr. (200 degrees Cent.), and also above 780 degrees Fahr. (415 degrees Cent.), while between these two temperatures, austenite is deformed without any phase change. These facts made it possible to determine that in dry grinding the work was deformed to a depth of about 0.0005 inch but that the temperature did not exceed 390 degrees Fahr. below a depth of about 0.0002 inch. In wet grinding, these depths were cut roughly in half. Other and slower methods of finishing the surface that were tried, including honing, sandpapering,

lapping, metallographic polishing, and superfinishing, resulted in a deformation depth of the order of 0.00005 inch, and the surface itself appeared to stay below 390 degrees Fahr.

The fact that the stainless steel was deformed by dry grinding to about the same depth as found for hardenable steels indicates that the other depth values quoted for stainless are also likely to hold for hardened tool steel, at least insofar as order of magnitude is concerned. This conclusion is further strengthened by the results of some auxiliary experiments performed by Wulff on beta brass, beta bronze, and austenitic manganese steel. In all these metals, metallographic examination showed the deformation effects of grinding to extend to a depth of about 0.0005 inch; moreover, there was evidence of less conclusive character that other methods of surface preparation resulted in deformation depths corresponding to those originally found for the stainless steel.

Metallurgical Factors

Hardening and Tempering Transformations—The susceptibility of a hardened steel to cracking during or after grinding depends primarily upon its metallurgical condition, and the same is true of its susceptibility to being left stressed after grinding. An example is furnished by the work carried out by Sweetser (3), who investigated the effects of extremely poor heat treatment of several different tool steels. He found that the more abusively a steel was heat treated, whether by too high or too low a hardening temperature or by the omission of tempering, the more gently it had to be ground to avoid the development of etch cracks upon subsequent etching in hot sulphuric acid.

To understand why this is so, it is necessary to consider certain microstructural characteristics of steel. It is now a well established fact that some austenite may be retained when a tool steel is hardened, the amount increasing with the alloying content. If the hardening operation has been properly carried out, either no austenite is retained or it can be eliminated by suitable tempering; however, improper hardening may make the austenite too stable to be broken down by the usual tempering procedures. The transformation product of austenite below the martensite point is untempered martensite, which has also been variously called white, tetragonal or alpha martensite. Steel in this condition has the lowest density and is exceed-

ingly brittle. At comparatively low tempering temperatures, this form of martensite undergoes progressive transformation with a slight decrease in volume to tempered or cubic martensite, also known as the dark or beta variety. This is practically as hard but not nearly as brittle.

A number of examples are available, in addition to the one already mentioned, to show how undesirable, from the standpoint of grinding, are retained austenite and untempered martensite. These two constituents can best be considered together because they are likely to occur simultaneously.

Working with a plain carbon tool steel, Monma (13) was able to eliminate spontaneous cracking of the ground surface by a low-temperature draw prior to grinding, which sufficed to change the untempered form of martensite into the tempered form. Brophy (2) observed essentially the same thing for etch cracks in steel of a similar composition.

d'Arcambal (21) has stated that high speed steel tools drawn before reaching room temperature are very susceptible to cracking when ground, the reason being that in such a case there is considerable retained austenite present. According to Morrison (8), a double draw, which is known to help in breaking down retained austenite and in tempering the resultant martensite, is effective in reducing the tendency of "sensitive" heats of tool steels to crack on grinding.

Along these same lines, some experiments carried out by the writer have shown the beneficial effect of properly tempering 18-4-1 high speed steel prior to grinding so as to keep high stresses from appearing in the surface after grinding. With all the other conditions kept the same, the hardened but untempered pieces had the most etch cracks; those tempered at 500 degrees Fahr. (260 degrees Cent.) had somewhat fewer etch cracks, with a further slight decrease for the pieces tempered at 800 degrees Fahr. (430 degrees Cent.); and the pieces that were properly tempered at 1050 degrees Fahr. (570 degrees Cent.) showed no etch cracks at all. It is obvious from the temperatures involved that microstructural changes were required, in addition to any stress-relieving that might have occurred, in order to keep down to a negligible level the surface stresses left after subsequent grinding.

Berlien (20) has found that cracking of ground surfaces may be eliminated in certain steels by subjecting them to sub-zero refrigeration for the purpose of completing the transformation of aus-

tenite, and then tempering the resultant martensite in the usual fashion before grinding. In this connection, it is interesting to note the report (22) that sub-zero refrigeration immediately after grinding may result in the formation of cracks. This could happen if a piece containing retained austenite was badly stressed by improper grinding so that these stresses, in combination with sub-zero transformation stresses, would exceed the low-temperature strength of the steel, in which the freshly formed martensite would be untempered and hence very brittle. The more logical procedure would be both to transform the austenite at sub-zero temperatures and to temper the resulting martensite before grinding, and not after.

These examples show clearly that retained austenite and untempered martensite should, when possible, be avoided in work that is to be ground but the mechanisms by which they exert their deleterious effects have not been established. It is possible that the inherent brittleness of untempered martensite will lead to cracks during grinding, or it may be that the heat of grinding will temper this martensite with a slight reduction in volume, resulting in tension stresses and perhaps in cracks. One function of the retained austenite may be to set up high stress concentrations as a result of its transformation to martensite, which would probably be accelerated by the grinding operation; whether such stress concentrations would help or hinder the formation of subsequent etch cracks is not known. It is possible, of course, that retained austenite would serve only as a continuous source of untempered martensite. Considerable work will have to be done before it can be decided just how these undesirable constituents make it more difficult to grind steel without cracking it or stressing it dangerously.

The evidence that cracking may be associated with the presence of untempered martensite does not invalidate the previously described hypothesis that tension stresses are set up as a result of the plastic deformation caused by the heat of grinding, since the martensite explanation cannot be used except for steel. Such cracks form in hardened steel, not only for microstructural reasons but also because of the purely thermal stresses, the two mechanisms acting in conjunction; otherwise, it would be impossible to explain why the cracks bear such a definite relationship to the grinding direction.

Elimination of the retained austenite and untempered martensite from hardened steel by proper heat treatment is only part of the problem since if grinding conditions are such as to generate excessive

heat in the work, fresh austenite may form, which, on cooling extremely rapidly because of the strong quenching effect of the adjacent cold steel, will change to untempered martensite. Metallurgical changes resulting from grinding heat are best dealt with as part of the general subject of burn.

Burn—The appearance of burn in a ground surface is an indication that moderately high temperatures were reached in grinding, resulting in a slight oxidation of the surface. Since the color of the oxidized film depends on both temperature and time, a higher temperature is required for a given color to form when it is caused by grinding heat, which acts for only an extremely short time at a given spot, than when it is caused by ordinary tempering. Based on the extensive study by McAdam and Geil (30) of the temperature and time relationships involved in the formation of oxide films on steel, a reasonable estimate of the momentary temperature needed to develop a slight burn in, say, a hundredth of a second on carbon tool steel is 750 degrees Fahr. (400 degrees Cent.). Steels more resistant to oxidation, such as some of the alloy tool steels, probably require an even higher temperature for slight burn to appear.

When hardened steel is burned in grinding, the temperatures reached may be high enough to cause some softening of the surface region, resulting in the well-known phenomenon of soft skin. Of course, this softening can also take place in the absence of visible evidence of burn, which is often cleaned off in the final passes. The extent to which tempering occurs at a point just below the ground surface of hardened steel depends on the time-temperature relationship for that point, and this in turn depends on the surface temperature and on the distance of the point below the surface. Ordinary grinding conditions may lead to a slight softening at the very surface while more severe ones may cause both a greater drop in hardness at the surface and a greater depth of softening.

This softening is one form of what may conveniently be termed metallurgical burn to distinguish it from visible burn, which has to do with surface oxidation. Metallurgical burn can be observed by etching with nital a metallographically polished section perpendicular to the ground surface. Surface stresses may accompany either kind of burn.

More severe grinding than that which simply softens the surface of hardened steel will lead to other varieties of metallurgical burn, besides making the visible burn more pronounced. Thus, there may

be occasional flashes of heat capable of rehardening the steel in spots. When such a surface is etched with 5 or 10 per cent nital, white specks of freshly formed martensite appear against a background of dark etching, thoroughly tempered martensite. With a further increase in grinding severity, a continuous layer of white martensite may form on the surface, the result being a moderately soft layer produced by tempering, which is located between the unaffected hard core and a thin superficial layer of martensite. The intermediate layer merges gradually with the hard core but there is a sharp line of demarcation between it and the new martensite on the surface. More complicated burn patterns can be observed when the martensite layer that was formed during one contact with the wheel is tempered and partly rehardened during a succeeding contact of sufficient severity.

The depth of the martensite layer depends on the grinding conditions and on the steel. Probably the most severe form of grinding is that which takes place when a cut-off wheel is used. A martensite layer 0.05 inch deep has been observed on the end of a bar of S.A.E. 1040 steel as a result of a wet cutting-off operation that was performed under unsatisfactory conditions. Even in commercial precision grinding operations, it is possible to grind severely enough to get a martensite layer 0.004 inch deep (9). Under less abusive grinding conditions, however, any martensite that is formed may be much too thin to detect by the usual methods.

A most unusual case has been reported by Murdock (4), in which grinding was employed to harden the surface of a steel punch without distorting it as happened when the regular heat treating procedure was followed. The grinding wheel was so conditioned as to create a great deal of frictional heat. Although this method was successful in this particular case, and no trouble from cracking was reported, the writer would hesitate to recommend grinding as a substitute for the more customary heat treating methods.

One of the most important characteristics of the heat treatments resulting from grinding is the extremely short duration, leaving out of the picture the slight tempering that may take place if the whole piece is given a chance to heat up. In this respect, the time intervals are many times shorter than those common in induction hardening or in flame hardening. A further difference between heat treating inadvertently caused by grinding and the usual type of heat treatment is that the former takes place at the same time as the stressing and

cutting operations and this too may affect the cracking tendency. A possible consequence of the very short time in which grinding heat has a chance to affect the steel is that the microstructure will not correspond to the highest temperature actually reached, just as Ellis (31) found that induction-hardened pearlitic steel could remain pearlitic in appearance although fully hardened to martensite.

Cementite—Carburized steel may be subject to cracks for an entirely different reason from that which has already been discussed. The heat treatment may be such that there is a considerable amount of thick grain-boundary cementite as a result of insufficient diffusion of carbon into the interior of the steel. Since every surface grain has the surface for part of its grain boundary, it follows that in the extreme case the whole surface could consist of grain-boundary cementite, which, being extremely brittle, would be very susceptible to cracking by grinding. In actual practice, cracks are sometimes found after grinding even in unhardened carburized steel, the reason being the presence of cementite concentrations on the surface. Similar cracks can, of course, occur in carburized steel if it is ground after hardening.

One way of preventing such cracks is to eliminate the brittle cementite surface by changing the heat treating procedure to permit the carbon to diffuse into the interior. With the carbon content near the surface brought closer to the eutectoid composition, there will be less chance of finding grain-boundary cementite. This means that although a properly carburized and hardened steel could still be cracked by grinding, its susceptibility to cracking under the same grinding conditions would be no greater than that of a tool steel having roughly the same composition as the carburized surface.

The deleterious effect of grain-boundary cementite is not restricted to carburized steel. For example, it can occur in plain carbon tool steel and it is interesting to note that Brophy (2) found etch cracks to form more readily in annealed and hardened steel than in the same steel when it was normalized prior to hardening. The explanation given was that a cementite network was present in the annealed specimens but not in the normalized ones.

Alloying Elements—It has often been found in practice that the highly alloyed steels are more difficult to grind without cracking or unduly stressing them than those low in alloying elements. The reason for this is not the presence of the alloying elements as such but the effect of these elements in preventing the completion of the hard-

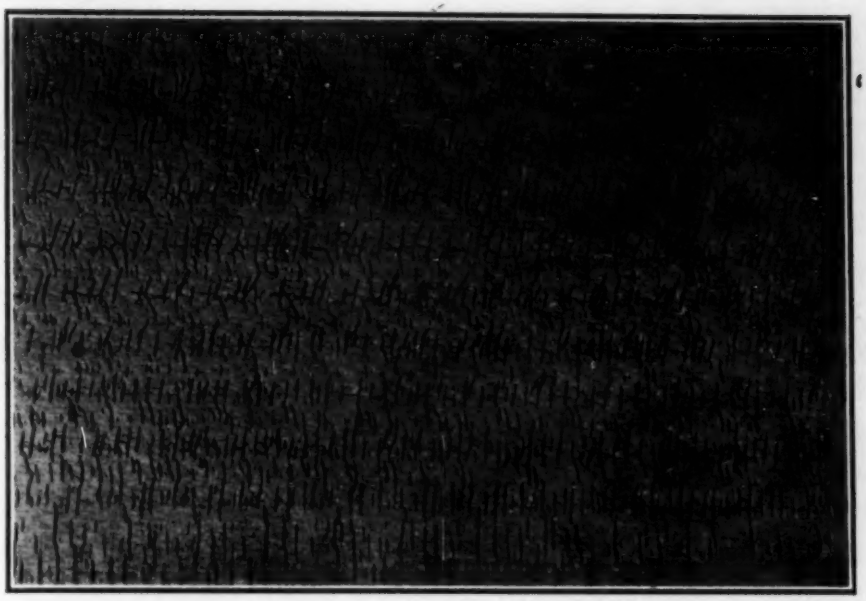

Fig. 11—Etch Cracks in Hardened 18-4-3 High Speed Steel. A hot sulphuric acid etch was required to develop these cracks from the stresses introduced by very severe grinding, which burned the work badly. No cracks were present prior to etching, nor were any etch cracks developed during prolonged immersion in cold acid, showing that the stresses in the surface were relatively low. × 2.

ening and tempering transformations under ordinary heat treating conditions. When care is taken to allow the hardening transformation to go to completion and to follow it with the proper tempering treatment, experience has shown that highly alloyed tool steels are no more likely to be cracked in grinding than is plain carbon tool steel; however, when the heat treatment is not sufficiently good, an increase in the alloying content is very likely to increase the incompleteness of the transformation and this makes it more difficult to grind without injury to the surface. The microstructure and hardness are the determining factors, and the chemistry of the steel is important primarily insofar as it affects the microstructure.

An illustration of this was furnished by a piece of properly hardened 18-4-3 high-speed steel, Rockwell C-65, which was ground so severely that the surface was very badly burned. However, no cracks could be detected in the ground surface nor could any cracks be developed by etching the piece in cold sulphuric acid for as long as 45 minutes. Other experiments have shown that a 15-minute etch in cold acid is sufficient to develop etch cracks in improperly tempered high speed steel that has been severely stressed in the grinding operation. Thus the absence of etch cracks after the cold etch in the piece under consideration indicated the absence of high surface stresses.

A subsequent etch in hot sulphuric acid developed the comparatively *low* surface stresses that must have been present into the fine and closely spaced etch cracks that can be seen in Fig. 11.

The same severe grinding conditions applied to pieces of oil hardening tool steel, such as those in Fig. 10, resulted in high surface stresses that showed up as etch cracks after only a 15-minute immersion in cold acid. Even though the two steels may have a somewhat different response to the etchant, it is clear that the 18-4-3 high-speed steel had a less severely stressed surface after grinding than did the oil hardening tool steel. Had the latter been hardened and tempered as satisfactorily as was the high speed steel, it is very likely that the susceptibility to etch cracking would have been more nearly the same for both. This example should dispose of the prevalent idea that highly alloyed steels are necessarily more difficult to grind without injury than those low in alloying elements.

Other Factors—It is generally true that the harder a steel is, the more likely it is to become cracked or left highly stressed after severe grinding. Sometimes, a few points decrease in the Rockwell C value of a hardened steel can change it from one that has to be ground very carefully to avoid cracking to one that cannot be cracked even under very severe grinding conditions. In some special cases, however, the microstructure is more important than the hardness. For example, high speed steel tempered at 500 degrees Fahr. (260 degrees Cent.) is several points softer than one properly tempered at 1050 degrees Fahr. (570 degrees Cent.), the increase in hardness being due to carbide precipitation; however, as has already been mentioned, the former will have to be ground more gently than the latter to avoid cracks or high surface stresses because of the presence of untempered martensite.

It is probable that for each steel there is a minimum hardness that has to be exceeded before cracks can be formed by abusive grinding. Ordinarily, this minimum hardness is in the vicinity of Rockwell C-60, but in certain special cases cracks have been found in steels no harder than Rockwell C-45. Experience indicates that cracks in these comparatively soft steels can occur only under conditions of broad contact between wheel and work, such as takes place in the form grinding of gears.

The presence of nonmetallic inclusions in steel has been shown by Brophy (2) to increase the likelihood of cracking. He found that dirty steel cracked readily under the same grinding conditions

that did not cause any cracking in clean steel of essentially the same composition.

Occasionally, certain lots of steel are found to be unusually sensitive to cracking, yet nothing can be found in their metallurgical condition, using ordinary test methods, to distinguish them from all the other lots of the same steel, which behave normally in grinding, i. e., do not crack easily. The most convenient way of dealing with these special cases in our present state of ignorance is to call them "sensitive steels" and let it go at that; future research will probably reveal the cause of this sensitivity to be connected with the hardening transformations.

Hard Nonferrous Alloys—A comparison of the susceptibilities to surface cracking after grinding of several hard alloys of a particular type shows that the metallurgical condition very likely does affect the susceptibility, but nothing much is known beyond this. For example, it is well known that the harder grades of cemented carbide, which contain less cobalt binder and hence are less ductile, are more likely to be cracked by improper grinding than are the softer grades. In special cases, it is possible to make alloys so brittle that surface cracking occurs even under the gentlest grinding conditions.

Grinding Factors

General Considerations—The susceptibility of a ground surface to injury by grinding depends both on the nature of the material, which we have discussed at length, and on the conditions under which the grinding is done. As anyone experienced in grinding realizes, there are innumerable sets of conditions, and the only way of dealing with them intelligently is to consider individually the effect of each type of condition on possible grinding injury.

The nature of the grinding operation determines the area of contact between wheel and work. It is an observed fact that increased area of contact increases the possibility of injury to the surface because the lessened wheel pressure results in duller wheel action. The contact may be essentially point contact as in ball grinding; or, at the other extreme, it may be complete contact over a large area, as when a flat surface is ground with the side of a flat wheel. In between are various kinds of approximately linear contact, this degree of contact being least for cylindrical grinding, intermediate for surface grinding and greatest for internal grinding.

In general, the contact area is automatically allowed for in the selection of the proper wheel specification for the operation in question, the initial selection usually being based on past experience; sometimes, however, the contact area changes sufficiently in a given operation to lead to a condition of injury to the work from a condition of no injury, as when, for example, a large wheel is substituted for a small one in internal grinding.

The wheel specifications include the kind of abrasive, its grit size, and the grade,[2] structure number,[3] and bond in the wheel. Each one of these may have an appreciable effect on grinding injury but we shall confine our present attention to the grade of the wheel as this is the principal variable with which the wheel user is likely to experiment in an effort to get the most satisfactory results under his particular grinding conditions.

Operating factors that we shall consider in some detail comprise the feeds and speeds. In the case of surface grinding with the periphery of a straight wheel on a reciprocating type of grinder, the variables are downfeed, crossfeed, traverse speed and wheel speed. Still other factors that call for comment in varying degree, depending on how much experimental evidence is available, are condition of the wheel face, method of cooling the work, and condition of the machine.

Experimental Conditions—The experimental data forming the basis of the subsequent discussion were obtained by surface grinding pieces of manganese oil hardening tool steel, hardened and tempered to Rockwell C-62 or 63, such as are shown in Figs. 2 and 10. The rounded shape of the specimens was due to the fact that they were cut from hardened rings that happened to be available in quantity. Diametrically opposite pieces from a ring were often used in duplicate to furnish a check on the accuracy of the data. With only occasional exceptions, the eight pieces cut from a single ring were found to be equally susceptible to etch cracks and the different rings used in these experiments also had very nearly the same susceptibility.

The grinding was done dry on a 6 by 18-inch Norton surface

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[2]The grade letter of a particular type of grinding wheel is a relative measure of the strength with which the abrasive particles are held together by the bond. Hardness or softness refers to this bonding strength and not to the hardness of the abrasive. The abrasive particles are held more firmly in a hard wheel than in a soft one, do not come out as readily after being dulled by use, and hence the grinding action is more severe as shown by the greater injury to the work. In the standard marking system, hardness increases according to the alphabetical order of the grade letter.

[3]The structure number of a grinding wheel indicates the relative spacing of its abrasive grains, and in a Norton wheel is related to the volume percentage of abrasive that is present. The higher the number, the wider is the grain spacing.

grinder equipped with a horizontal spindle. This machine has a reciprocating table motion. Constant grinding conditions were used throughout, except for the variations of the one condition whose effect was being studied in a particular experiment. The conditions were:

Wheel	Norton 38 Alundum, 38A46-J8VBE
Wheel dimensions	7 to 8 inch diameter, ½ inch width
Wheel dressing	Moderately sharp, using a sharp diamond
Wheel speed	6000 surface feet per minute
Crossfeed	0.050 inch per pass, direction of crossfeed reversed after each completed crossfeed
Downfeed	0.002 inch per completed crossfeed
Total downfeed	0.020 inch
Traverse speed	40 feet per minute

In one case, regular Alundum wheels were used in place of 38 Alundum wheels.

These standard conditions were severe enough to stress the pieces sufficiently to cause the appearance of a reasonably large number of etch cracks after a 15-minute etch in cold 15 per cent sulphuric acid, which had been treated with ferrous sulphate as previously described.

It is to be noted that the amount of etch cracking was used as the measure of injury to the ground surfaces, this being done because it was found more convenient, from the standpoint of reproducibility of data, to work with etch cracks rather than with cracks spontaneously generated in grinding; however, the remarks apply equally well to the spontaneous cracks, which would have been obtained, with poorer reproducibility, under more severe grinding conditions. As previously pointed out, an increase in the amount of etch cracking under constant etching conditions corresponds to more severe surface stresses.

Effects of Wheel Variables—One of the most elementary facts about fixed-feed grinding is that the harder the grade of the grinding wheel, the more likely it is to injure the ground surface. Since wheel wear decreases as the grade of the wheel is made harder, proper grinding practice customarily calls for the use of the hardest wheel that will not injure the work. An example of the effect of wheel grade upon the stresses introduced into the ground surface is illustrated in Fig. 12. The bottom curve is concerned with the etch cracks corresponding to the highest stresses only, while the top curve shows how additional etch cracks develop from lower stresses. Even harder wheels would have to be used under the conditions of the experiment

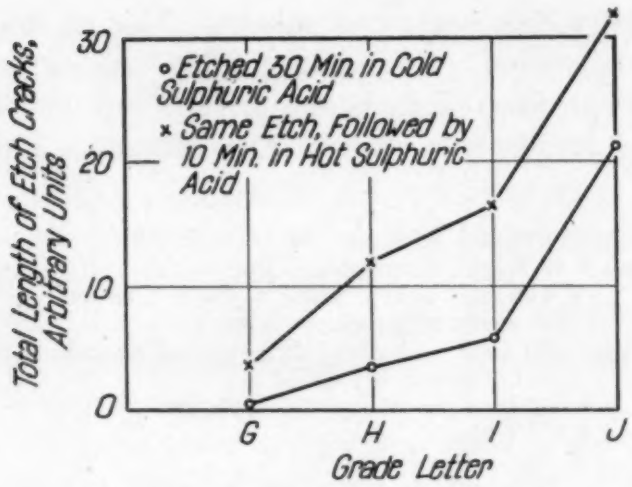

Fig. 12—Effect of Wheel Grade on Stresses Introduced Into Surface by Grinding. Specimens like those in Fig. 10. Norton Alundum wheels, 46 grit size, were used in this experiment. Purpose of the cold etch was to develop etch cracks from the high stresses; of subsequent hot etch, to bring out the lower stresses in addition.

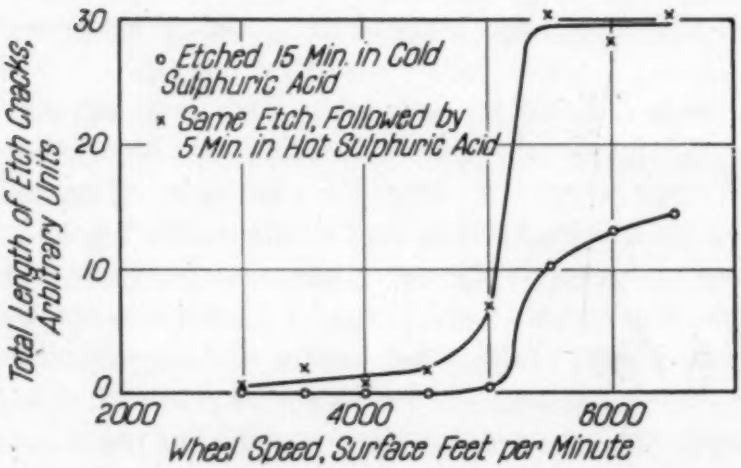

Fig. 13—Effect of Wheel Speed on Stresses Introduced Into Surface by Grinding. Specimens like those in Fig. 10. Norton 38 Alundum wheel, 46 grit size, K grade, was used. Cold and hot etches used to develop high and low stresses, respectively, into etch cracks.

to raise the stress level sufficiently to form spontaneous cracks in the surface during grinding.

In this, as in other cases, it has been found convenient to use the total length of the cracks in the specimen as a measure of the surface stresses, the length being expressed in arbitrary units equal to the width of the specimen. Experience has shown that the number of cracks is not as satisfactory a criterion as their total length.

The effect of wheel speed upon surface stressing is shown in a

similar plot in Fig. 13. Here we see in the lower curve that high stresses begin to appear only after the wheel speed exceeds approximately 5000 surface feet per minute. The upper curve, obtained by an additional etch of the same specimens in hot acid to bring out lower stresses, shows that these are present to a slight extent at even the lowest speeds tried and to a greater extent at the higher speeds. The considerable difference between the right-hand portions of the two curves can best be interpreted as being due to the presence of stresses in the ground surface ranging in magnitude from just below the high ones, which had already been developed into etch cracks by the first etch, to the lowest stresses that could be developed by the hot acid etch.

Long experience has shown that grinding conditions corresponding to the left-hand portions of the curves are not at all injurious to the work under ordinary conditions of usage. Thus the presence of a few etch cracks following a hot acid etch should not be taken as a sign of injury by stressing, for the stresses are too small in magnitude to be worth considering. It is a different matter, of course, when numerous etch cracks are brought out by a brief cold etch since these indicate that stresses are present after grinding which are almost high enough to crack the surface region spontaneously.

Another important factor is the condition of the wheel face. This depends at first on how the wheel was dressed. In one experiment it was found that turning the diamond a little in its holder so as to change its effective sharpness in dressing the wheel caused sufficient increase in the wheel sharpness to keep any etch cracks from forming where previously they had been numerous.

When the effect of the dressing has worn off as the result of a certain amount of grinding, then the condition of the wheel face is dependent upon the grinding operation itself. An illustration of this is furnished by an experiment in which twenty-four specimens were ground in succession after the wheel had first been dressed fairly sharp with a diamond. The first few pieces to be ground were found to have short straight etch cracks, a typical specimen being A of Fig. 10. Here the grinding took place only with the leading edge of the wheel. As more pieces were ground, the wheel wore and the area of grinding contact gradually increased until the last pieces were ground with the leading half of the wheel periphery. The etch cracks correspondingly lengthened (B), branched out (C), and started to form a network (D).

In this example, the increased area of contact reduced the load on individual abrasive grains, permitting them to become dull. This caused both an increase in the amount of etch cracking, which has already been discussed in connection with wheel sharpness, and an increase in the complexity of the pattern. Other experiments have shown that increased area of contact is primarily responsible for increased complexity. This is to be expected since the stresses in a wide area that has been suddenly heated and cooled should be more complicated than those in a very narrow area.

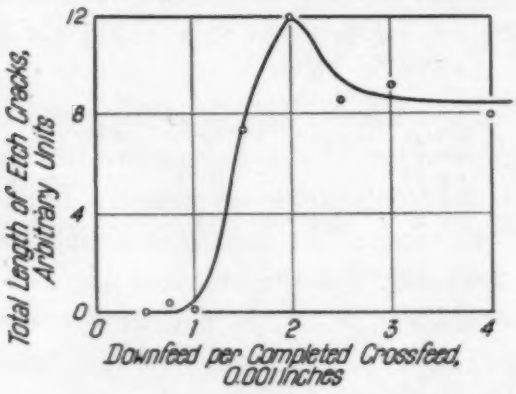

Fig. 14—Effect of Downfeed on Stresses Introduced Into Surface by Grinding. Specimens like those in Fig. 10, etched in cold sulphuric acid to develop high stresses into etch cracks.

The shape and size of the surface being ground may have an important bearing on the condition of the wheel face. When small pieces are being ground, the repeated contact with their sharp edges will tend to keep the wheel sharp; on the other hand, continued contact with the surface of a large piece will tend to dull the wheel. Thus we may find that a large piece is stressed more severely by grinding than a small one although all the grinding conditions are nominally kept the same.

Effect of Machine Variables—One of the simplest ways of changing the severity of surface grinding is to vary the downfeed. The effect of increasing the downfeed per completed crossfeed is illustrated in Fig. 14. It is to be noted that there is a maximum in the amount of etch cracking at the 0.002-inch value, beyond which the curve drops somewhat and then stays constant. This sort of behavior might appear to contradict the idea that increased grinding severity leads to more serious stressing, but there is good reason to believe that the grinding can be more severe at some intermediate

downfeed and not necessarily at the highest one. At this intermediate downfeed, the wheel becomes progressively duller during grinding because the pressure acting on it is insufficient to break it down and remove the dull grains. At higher downfeeds, the pressure is great enough to break down the wheel as it begins to dull in use, and as a result of its staying sharp, the severity of grinding may thus be less than for the intermediate downfeed. This explanation also holds for the fact that a light feed may burn the work when a heavy one will not. Of course, the heavy feed results in more wheel wear.

The effects of crossfeed and traverse speed have not been investigated in any detail. It is obvious, however, that an increase in crossfeed corresponds in general to an increase in grinding severity. In several experiments, much more etch cracking was observed for a 0.050-inch crossfeed than for a 0.025-inch crossfeed. As regards traverse speed, it is known that increasing it may aid in keeping the wheel sharp, and sometimes in practice such an increase reduces the occurrence of cracking during grinding.

A factor that deserves some consideration is the effect of the coolant upon grinding severity. Usually, wet grinding properly carried out is less likely to result in injury to work surface than is dry grinding. However, insufficient flow of coolant or improperly directed coolant can be worse than no coolant at all. With such poor coolant practice the work is permitted to become fairly hot in some spots, which are then suddenly quenched; in dry grinding, although the surface becomes even hotter, it will cool more uniformly. When coolant is used, care should be exercised to see that the work is flooded with it.

Even when there is enough coolant flow to flood the work, there may be unusual cases in which dry grinding is more satisfactory. One such case was that of an alloy cast iron containing a large amount of chromium. Dry grinding made it possible to use a wheel two grades harder than could be used in wet grinding if spontaneous surface cracks were to be avoided. It was noted that the work was burned when ground wet and not burned when ground dry with the same wheel; perhaps the coolant caused the wheel to become dull in use, leading to the observed burning and cracking.

Beneficial effects are often associated with the use of a straight grinding oil free from any water, in comparison with water coolants containing a soluble compound in dilute solution such as, for example, a 40 to 1 dilution. Experience has shown that a wheel that

becomes dull when water coolants are used is more likely to stay sharp with a straight grinding oil.

In one case known to the writer, occasional lots of hardened high speed steel rods would crack when rough-ground using a soluble oil solution for coolant but the trouble disappeared when a change was made to straight grinding oil. Although the work heated up considerably when grinding oil was used, this being a poor coolant, yet the continuing sharpness of the wheel kept the grinding from becoming severe enough to cause cracks; in contrast, grinding with soluble oil solution stressed the surface severely enough for cracks to appear even though the work as a whole remained cold. The strong quenching effect of the soluble oil solution, compared to the mild effect of the grinding oil, may also have contributed to the crack formation. Similar benefits have been reported in the grinding of Alnico when grinding oil was substituted for soluble oil solution.

The condition of the machine may also be important from the standpoint of grinding injury. A machine with loose or worn working parts is likely to set up vibrations that will result in occasional spots being more severely ground than the rest of the surface, and high stresses or even cracks may appear at such spots. The poor condition of the machine may also be reflected in the condition of the wheel face and thus lead to poor results indirectly.

The last factor to be considered is the human element. The skill of the operator may be very important for it is sometimes observed when grinding is done on a production basis that one operator has little if any trouble from cracking while others encounter it continually. The method of payment for the work may determine the prevalence of cracking for it is generally true that on the basis of over-all economy a worker paid by the piece requires a harder, more durable grade of wheel than one paid by the hour. Both the harder grade wheel and the tendency to higher production may lead the piece-worker to have more cracking trouble than the hourly worker. These, however, are essentially problems of management and, as such, are beyond the scope of this paper.

Combined Effects of Metallurgical and Grinding Factors

We can now profitably turn to a brief consideration of how the various forms of injury to the ground surface of hardened steel are affected by the simultaneous action of the metallurgical condition and the grinding severity.

In order that the work be susceptible to being cracked during grinding, it has to be left in a sufficiently hard condition by the heat treatment; moreover, the degree of completeness of the hardening operation and the tempering treatment will both affect the severity of grinding necessary to crack the work. In other words, the more improperly a piece of steel is heat treated, the more gently it has to be ground to avoid cracks.

For a given metallurgical condition, the time at which surface cracks appear can serve as a rough index of grinding severity. The most severe grinding leads to the formation of cracks during the grinding operation itself or immediately thereafter. As the grinding is made progressively less severe, there comes a point at which the cracks begin to appear only after considerable cooling has occurred, assuming dry grinding. Here the stresses resulting from grinding have to be augmented by cooling stresses if cracks are to appear, and the cooling rate may be the determining factor as to the occurrence of cracks.

A further reduction in the severity of grinding may delay the initial appearance of any surface cracks until some time after cooling has been completed. Such cracks are sometimes observed to form only a few minutes after grinding has ceased and the work has cooled while in other cases the delay may be a matter of hours. In all instances of delayed cracking, it is reasonable to assume that stresses resulting from delayed microstructural changes, such as that of untempered martensite to tempered martensite, are gradually added to the surface stresses introduced during grinding until the total stress at some point becomes great enough to form a crack. It is likely that stresses introduced into the work surface during grinding serve to initiate or aid these microstructural changes that would otherwise be unable to occur.

These delayed cracks will appear as etch cracks in hardened steel if the work is etched suitably before any cracks have been given a chance to develop spontaneously. Here the distinction between etch cracks and spontaneous cracks is hazy because the name to be applied depends on whether high stresses in the ground surface are supplemented to the required extent by stresses from the etching procedure or by stresses gradually developed internally, presumably as a result of delayed microstructural changes. This lack of any clear-cut boundary between the two kinds of cracks shows that the only essential difference between the stresses giving rise to them is that of

magnitude, and even in this respect there is an overlapping range depending upon the time that has elapsed between grinding and etching.

As stresses introduced by grinding are made lower by decreasing its severity, a corresponding rise in the severity of the etching procedure is required if etch cracks are to appear. This, as we have seen, can be accomplished by increasing the etching time or the temperature of the etchant. When the stresses are low enough, it becomes impossible to produce etch cracks no matter how prolonged the etching treatment may be or how high the etchant temperature. However, the presence of such stresses can be demonstrated by the warping of a thin strip after it has been ground, as described previously. While surface cracks caused by grinding are limited to a few types of metal, and etch cracks have so far been produced only in hardened steel, the phenomenon of warpage is not restricted in this manner.

It is also possible to vary the metallurgical condition of the steel while keeping the grinding conditions constant, and a similar sequence of crack types will result. Thus where an untempered hardened steel is found to be cracked during the grinding operation, a slightly tempered version of the same steel may be found to be cracked only after some delay, while more thorough tempering may limit the stressing sufficiently to keep spontaneous cracks from forming at all, so that only etch cracks can be developed.

In so far as burn is concerned, its occurrence depends almost entirely upon the grinding conditions. A moderate change in hardness caused by varying the tempering procedure for hardened steel has practically no effect on burn. Of course, if the hardened steel were completely softened, its susceptibility to burn might be changed appreciably, but such a pronounced change in hardness properly calls for a corresponding change in the grinding conditions, including wheel specification.

A practical example of this, in which the soft condition of the steel is brought about by decarburization rather than by annealing, is in the grinding of hardened gear teeth. Unless the decarburized surface is entirely removed after heat treatment by thorough blast cleaning, grinding may burn the teeth because the soft surface has a tendency to load and dull the wheel rapidly, thereby generating sufficient heat to burn the steel. This condition of the wheel face may remain even after all the decarburized layer has been ground off and in this case the burn would continue to appear on the hardened steel,

although if the wheel were in properly sharp condition, burn would not now occur.

Possible Effects in Use of Injured Ground Surfaces

There is little that need be said about the undesirability of surface cracks in ground parts, whatever the principal cause of the cracks may be. As we have seen, cracks can appear after grinding only if the material is hard enough and this usually means that the part will be put to some type of service in which cracks are highly objectionable. Cutting tools and gears, to cite two examples, are used under conditions of variable stress, which means that even tiny surface cracks are likely to propagate into the interior and greatly reduce the service life of the part.

Almen's work (18) has shown clearly that the fatigue properties of steel can be affected adversely by tension stresses in the surface, and that the same properties can be improved by stressing the surface in compression, as by shot peening. The possible effects of surface stresses in ground parts are much more difficult to evaluate because so little is known of the damage to fatigue properties caused by abusive grinding. What little is known has not been generally available in detail on account of security restrictions. In the great majority of applications, where the part is not subjected to extremely high alternating stresses in service, there is no evidence to indicate that grinding, properly done, has any deleterious effect on the useful life of the part.

What is likely to be harmful is the presence of high tensile surface stresses that can be easily developed into etch cracks, since a piece stressed to that extent can be subjected to but little external stress before the cracking point is reached. Such high stresses are also a danger signal because very slight and possibly undetectable changes in grinding conditions or in the condition of the material being ground may intensify the stresses sufficiently to cause cracks to appear spontaneously in similar pieces ground at a later time. As the magnitude of the tension stresses in a ground surface is decreased by any means whatever, the piece becomes capable of withstanding progressively more severe usage. Just what the relationship may be between stress level in the ground surface as indicated by the amount of etch cracking, and the external stresses that can be safely applied remains to be determined experimentally, under both static and fatigue conditions.

The presence of surface stresses may also be harmful where even a minute amount of distortion is not permissible, the warpage of a thin strip being an extreme example of such an undesirable effect.

Burn, as we have seen in the section on metallurgical factors, can soften hardened steel, or if severe enough, it can even reharden the surface. Whether the softening of the surface caused by severe grinding is harmful depends on the circumstances. Soft skin, as it is often called, is a serious defect when it occurs on the working surface of a cutting tool since it greatly decreases the useful life of the tool. In some cases, soft skin of a harmful nature has gone undetected because the work surface did not appear to be burned, having been cleaned off in grinding. Its presence would have been revealed by a nital etch. On the other hand, there are many ground surfaces whose function is not impaired by a slight softening, yet they are sometimes rejected because too high an inspection standard has been set. In one such instance, a hardened and ground steel surface with no signs of burn showed dark spots after it had been etched with nital, yet a careful metallographic examination of a section through the region that was darkened on the surface failed to reveal any signs of darkening. This indicated that the tempered region was too shallow to be resolved by the microscope, and in this case, the softening of the surface could hardly be considered to be harmful.

There are other examples of surface-finishing operations that have resulted in visible changes in the surface metal without there being any reason to suspect that such operations injured the metal in any practical sense. Almen (18) found that shot peening brought out the machining marks in a surface that had been rough machined, carburized, hardened, ground, and polished, but stated that there was no evidence of any real harm to the surface. Wulff (15) observed that even lapping or honing visibly affected an extremely thin surface layer, and these finishing operations are generally conceded to be very gentle and free of injurious effects. Cases like these show that not all detectable changes in a ground surface are necessarily injurious to the work. The subject of which changes are permissible and which are harmful under particular conditions of use is one needing considerable investigation.

The relationship between visible burn and cracking deserves mention because there is a tendency to connect the two more closely than is really justified. Burn is sometimes taken as an indication of possible cracks resulting from improper grinding. There can be no

objection to this procedure in specific cases when it is based on experience but there is no good reason for extending it indiscriminately to all cases of burn. Cracks and burn are associated only to the extent that both are caused by grinding heat. In some cases, it is possible to get cracks in the absence of any visible burn, either because it did not form or because it was cleaned off; in others, the work can be burned very badly, yet no cracks result. This is because burn is essentially a function of the temperature to which the surface has been brought momentarily, while cracks depend for their formation primarily on the stress condition of the material ground.

When grinding conditions are sufficiently severe to result in a layer of freshly formed martensite in the surface of hardened steel, it is likely to become very highly stressed or to crack; however, if the steel is below a certain hardness level, the presence of such a martensite layer cannot cause any surface cracks to form nor can any etch cracks be developed. An instance in which the presence of this freshly formed martensite is most undesirable for its own sake is furnished by bar stock that has been improperly cut off with an abrasive wheel, resulting in the hardening of the cut ends. Subsequent machining of such pieces is likely to lead to tool breakage.

Another injurious effect seems to be connected with the presence of extremely small regions of white martensite in a ground surface of hardened steel. Although the mechanism is unknown at present, these specks of martensite seem to be associated with premature failure of the part under fatigue conditions even though there is no indication of any etch cracking after drastic etching of the part.

Some Methods of Preventing Surface Injury

A number of things can be done to eliminate the various forms of injury that may at times be associated in some way with ground surfaces, or to eliminate the harmful effects that may result from such injury. In previous sections, we have discussed defects present prior to grinding, the metallurgical factors that may be responsible for what superficially appear to be purely grinding troubles, and the numerous grinding factors that have to be kept in mind in considering severity of grinding. Some of the possible remedies, involving such things as improving the heat treatment in some respect, or decreasing the severity of grinding, have been indicated in a general way. When one of the factors resulting in injury to the ground sur-

face predominates over the rest, the remedy is likely to be easy to decide upon. In many cases, however, more than one factor is involved to a considerable extent, and then the decision as to what to do to eliminate the injury to the surface may depend on the economic aspects of making the various possible changes.

Another thing that can be done sometimes is to overcome the undesirable consequences of injury to the ground surface by a special treatment following the grinding operation. In hardened steel,

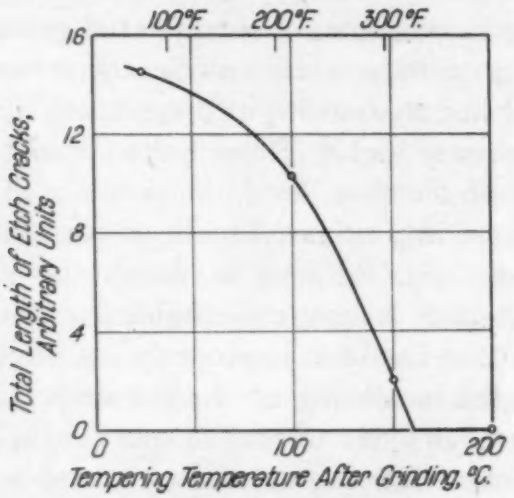

Fig. 15—Effect on Etch Cracks of 1-Hour Tempering Treatment Following Grinding. Specimens like those of Fig. 10, etched in cold sulphuric acid.

residual surface stresses can be diminished by tempering sufficiently, and thus the formation of etch cracks can be prevented no matter how severe the etching treatment used in the attempt to develop them. The temperature needed to eliminate completely the possibility of etch cracks depends on the steel. Thus the temperature for plain carbon tool steel (2) does not have to exceed 300 degrees Fahr. (150 degrees Cent.), while for an oil hardening tool steel (8) this temperature is insufficient and may have to be raised to 350 degrees Fahr. (180 degrees Cent.). Surface stresses in high speed steel, presumably 18-4-1, were partly relieved by tempering between 500 and 800 degrees Fahr. (260 and 430 degrees Cent.) but a temperature of 1050 degrees Fahr. (570 degrees Cent.) was required to prevent the appearance of any etch cracks upon severe etching (8). In all cases cited, the time at temperature was 1 hour. These temperatures suggest that the so-called stress-relieving treatments following heavy grinding are in reality treatments for tempering white martensite,

presumably freshly formed in an extremely thin layer by the heat of the grinding operation.

Some data are presented in Fig. 15 to show how tempering temperature following grinding affects the residual surface stresses in an oil hardening tool steel, as measured by the amount of etch cracking caused by a 15-minute immersion in cold sulphuric acid, which brings out the high stresses. One hour of tempering at a temperature somewhere above 320 degrees Fahr. (160 degrees Cent.) sufficed to drop the surface stresses to a level below that needed for the development of etch cracks.

These phenomena are of interest in another connection. They disprove Schrader's contention (5) that the occurrence of etch cracks is related to the surface finish and not to the stresses present in the surface. It is obvious that if the roughness of the surface were of any real importance, etch cracks should continue to appear after the low-temperature tempering treatments that have been discussed.

In the practical application of the tempering treatment, it has been found convenient to place the work in hot oil. Thus hardened and ground gears are often kept in oil at 350 degrees Fahr. for several hours to relieve stresses following rough grinding. Another practice following grinding that has seen some use, especially in connection with cast-iron surfaces, is to soak the piece for several hours in boiling water to which soda has been added as a rust preventive. This is done to relieve stresses that otherwise may distort the work at some later time.

Mechanical means are also available to decrease or eliminate undesirable surface stresses. It has already been shown that such stresses can be dealt with by very gently grinding away the stressed layer, which has generally been found to be about 0.0005 inch deep. The gentle grinding may, of course, leave some very low and undetectable grinding stresses, but these will normally be too low to be of any practical consequence. Honing could also be used to remove the layer stressed in tension and to leave low compressive stresses, as has been shown by Almen (18); lapping would also probably accomplish the same object.

Gentle grinding can often be used to remove cracks that have appeared spontaneously after heavy grinding, but this method cannot be counted upon with assurance since sometimes the cracks keep extending themselves below the surface as rapidly as the portion near the surface is ground away. Even if this method works, it is prac-

ticable only in special cases because the cracks may be deep enough to cause the work to be undersize after their removal.

Another method that may be useful in eliminating high surface tensile stresses involves shot peening the surface after rough grinding but before finish grinding. In this case, tension stresses in the surface present after grinding are overcome either partly or completely by the compression stresses introduced by shot peening.

When the necessary physical properties of the material are also responsible for great susceptibility to cracking as a result of grinding, all that can generally be done is to accept that fact and adjust grinding conditions accordingly in the direction of gentler grinding. With hardened steel, however, it is often possible to eliminate the trouble by modifying either the heat treatment or the grinding conditions. As we have seen, even if the *immediate* cause of spontaneous cracks in hardened and ground steel is grinding, the *basic* cause may be the heat treatment. For example, an improperly tempered tool steel may crack even when ground with care because its microstructure is such as to make it extremely susceptible to cracking. Here the real source of trouble is the undesirable microstructure, yet grinding is sometimes held to be at fault simply because the cracks appear only after grinding. Undoubtedly it would be possible in such a case to eliminate the cracks by changing to extremely gentle grinding conditions, but the proper solution is to improve the heat treatment.

It is much more difficult to decide what to do when only a few pieces in a lot are found to be cracked after grinding, or perhaps only one lot out of a number of lots. Small variations in grinding conditions or in the condition of the steel when it is ready for grinding may be responsible. An unfavorable combination of the two factors is also a possibility. Use of more uniform steel, improvement in the heat treatment or closer control of it, or less severe or more uniform grinding may eliminate the occasional cracks that do occur but the cost is almost certain to be raised, and it may be cheaper to permit a certain percentage of rejections on account of such cracks. This can be safely done if the magnetic-powder method is used for inspection purposes. The problem here is not to apportion the blame, but to find the least expensive solution, and obviously this type of problem cannot be settled independently by the heat treating department or by the grinding department, each trying to reduce its own costs.

The same considerations apply when high stresses in the ground

surface are to be avoided. In this case, proper stress relief after grinding is likely to be cheaper and more effective than changing the heat treating or grinding procedures.

Conclusions

Cracks, high stresses and burn are three possible types of injury that may be found in ground surfaces. In hardened steel, cracks and stresses may be the result of grinding under normally satisfactory conditions an improperly heat treated part, or of improperly grinding a correctly heat treated part, or of a combination of improper grinding and improper heat treatment. In other hard metals, like cemented carbides, where heat treatment is not a factor, any injury is attributed to improper grinding, unless it is an exceptional case of an alloy that, because of its brittleness, is extremely susceptible to surface cracking even when ground under the gentlest conditions possible.

Surface cracks are limited to hard metals, the hardness necessary before any cracks can be made to appear by abusive grinding depending on the kind of material, usually being about Rockwell C-60 for hardened tool steel; surface stresses and burn, on the other hand, may occur independently of hardness. Cracks are practically always injurious in that they are likely to propagate and cause failure; surface stresses and burn, on the other hand, may or may not be injurious, depending on the kind of service the ground part is expected to withstand. High tensile stresses that may be introduced by improper grinding will reduce the fatigue properties. Softening of the working surface of a cutting tool as a result of burn will shorten the life of the tool.

There is a close relationship between cracks and stresses in a ground surface because the former can occur only when the sum of the stresses, whatever their origin, exceeds the rupture strength of the material. How severely a piece can be ground without cracking depends on how high the stresses already present in the surface are relative to its rupture strength.

In hardened steel, improper heat treatment may either raise the level of the undesirable stresses or lower the rupture strength, the net effect in either case being less margin left for stresses that can be introduced by grinding without cracking the piece. Important metallurgical factors associated with cracks in ground work are re-

tained austenite, untempered martensite, and grain-boundary cementite. Both laboratory and practical studies have shown that the elimination of these constituents by suitable heat treatment reduces the susceptibility to cracking of a ground surface.

Grinding factors that may result in surface injury are too numerous to mention but they all involve excessive heat, either because of too much stock removal per unit of time or because of the frictional heat generated by dull abrasive grains in the grinding wheel. It is fairly certain that the magnitude of the stresses introduced into the surface by grinding increases with the severity of grinding although no quantitative stress data are available. Grinding conditions can practically always be improved to eliminate injury although in some cases this will involve an increase in the costs. When both metallurgical and grinding factors are involved, the question of to what extent each should be improved should be decided on the basis of over-all economy.

Cracks in a ground hardened steel surface can best be detected by the magnetic-powder method, if they are not directly visible. Dilute nitric acid can be used for the same purpose. Surface stresses, when they are above a threshold level, can be developed into etch cracks by etching with hot hydrochloric acid or with sulphuric acid, either hot or cold. Increasing the etching time or the etchant temperature makes it possible to develop lower stresses into etch cracks. The depth of surface cracks, occurring during or after grinding, is generally around 0.010 to 0.020 inch but the depth of the stressed layer from which such cracks originate appears to be no greater than 0.0005 inch.

Cracks in nonmagnetic hard materials, like cemented carbide, can best be detected by some method involving seepage of oil or similar substance from the cracks. Nothing is known about stress detection by etch cracking in hard metals other than hardened steel.

In spite of their comparative shallowness, cracks in a ground surface can rarely be removed in ordinary practice, either because of insufficient stock left for grinding below the crack bottoms or because the cracks keep propagating themselves deeper while they are being ground away. Surface stresses introduced by grinding, however, can be greatly diminished in magnitude in most hardened steels by tempering them at comparatively low temperatures.

Burn, which is the result solely of excessive grinding heat and does not depend on the previous stress condition of the ground sur-

face, can be eliminated only by a suitable improvement in grinding conditions. A burned surface is discolored by oxidation caused by momentary high temperatures. Even in the absence of this burn color, metallurgical changes may occur in the surface layer of severely ground steel. A soft skin on hardened steel is the result of momentary tempering somewhere below the lower transformation point while the presence of hard martensite shows that the temperature rose above this point. These metallurgical changes in hardened steel can be detected by a nital or nitric acid etch.

Acknowledgments

The author is indebted to H. W. Wagner and L. H. Milligan of the Norton Company Research Laboratories for their many valuable suggestions and criticisms in the preparation of this paper.

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DISCUSSION

Written Discussion: By F. B. Doane, president, Magnaflux Corp., Chicago.

The work of Dr. Tarasov in co-ordinating published information regarding the important class of defects known as grinding checks or grinding cracks, together with his field experiences and the result of his experimental work, is extremely timely and throws a great deal of light on a subject concerning which entirely too much misinformation exists. It is the purpose of this discussion to comment on certain conclusions of the author regarding methods of detection rather than on the causes and metallurgical nature of injuries in ground surfaces.

The author's estimation of the possibilities and limitations of the Magnaflux method are in keeping with the best knowledge of today. It is suggested that the term "magnetic particle inspection" be employed in place of the term "magnetic powder inspection", the former being considered more generally descriptive and also having been adopted by various code-making bodies such as the A.S.T.M. and others.

In searching for cracks in cemented carbides, a procedure has been worked out which has proven reasonably successful. This is based on the so-called Magnaglo, or fluorescent magnetic particle inspection. This method employs a light oil suspension of very fine ferromagnetic particles to which a fluoragent has been added, and the indications are viewed under "black light." In applying this procedure to cemented carbides there are two requirements. The first is that the carbide tips to be examined be separate and not brazed to the tool. This is due to the fact that the permeability of the material is very low and the resultant leakage fields are very weak, and are consequently seriously interfered with by strong leakage fields from the tool itself when present. The second requirement is that a strong and localized method of magnetization must be used and for this purpose yokes have been found to be most successful. The method has given satisfactory results on tips constructed of from 2.5 to 16 per cent cobalt, locating not only surface but subsurface defects as well. Most tips used in our work have been unground, but there is no apparent reason why the method should not be successful on ground tips also.

Magnetic particle inspection is not recommended for Alnico for the reason that too much difficulty is experienced with grain boundaries becoming strongly magnetized, strings of magnetized particles, called "whiskers", and local poles, due to inhomogeneity of the material.

For inspecting Alnico, Zyglo is preferable. A more general term for Zyglo is suggested which is "fluorescent penetrant inspection."

It is usually not considered necessary to soak cemented carbides in hot penetrant, and in this connection a revision of the author's technique is suggested. It may be that the cracks have other material in them such as water or coolant, if these were used. A usually satisfactory technique is to boil cemented carbide tips 10 minutes in a 10 per cent trisodium phosphate solution—rinse and dry—apply penetrant—drain 30 minutes—wash and dry and apply dry developing powder. Completed, the tools, brazed and ground, gen-

erally do not require boiling in trisodium phosphate. Some variations of the penetrant and development times may be required to meet special conditions or to establish the most efficient production inspection technique.

In concluding this discussion one more suggestion might be based on the use of magnetic particle inspection as a preventive control method in production. In practice this is done by periodically spot checking the output of each machine in the grinding room. A portable unit is useful for the purpose as it may be moved around the shop checking a few parts from each machine. As soon as any cracked parts are found the grinding procedure is checked and corrected.

Written Discussion: By Robert S. Moore, Research and Development Division, Quaker Chemical Products Corp., Conshohocken, Pa.

Dr. Tarasov's paper is indicative of the interest which the progressive abrasive companies are showing in the injuries which can be caused in grinding metal surfaces, and he is to be congratulated on his scholarly and yet practical approach to the subject. Let us hope that this is just the beginning of a series of studies in this field.

The effects, under usage conditions, of improper grinding which results in stresses of a lower order of magnitude than those which accelerate attack of etchants, but which are nevertheless mechanically important, are becoming well understood. These effects include undue wear and lowering of fatigue limits of ground components while in service. But methods of study and analysis of such stresses are embryonic. Gamma-radiography, Roentgenography, and flux density measurements all show interesting possibilities. Stresses not detectable by etch techniques but which, nevertheless, can be delineated by fatigue tests on samples ground under various abrasive, mechanical and medium conditions, indicate fertile fields of research ahead.

As Dr. Tarasov states, the question of the extent to which grinding factors should be improved in any instance should be decided on the basis of overall economy. It is altogether likely, however, that great strides can be taken in grinding efficiency with no unfavorable change in the economic aspects. This is particularly true in the grinding of hardened materials. Changing concepts of surface finish will aid in the more general utilization of the dull, satin surfaces which can often be obtained with greatly decreased subsurface stresses by grinding techniques using clean, free-cutting wheels having the most advantageous "load per grit".

Dr. Tarasov's account, under "Effect of Machine Variables", of an alloy cast iron which showed burns when ground wet but not when ground dry may be indicative of the deleterious effects of using a liquid medium conducive to loading or glazing.

The higher pressures required to break down (dress) a loaded wheel result in excessive heat.

The usefulness of a grinding wheel is limited by the ability of the abrasive to maintain a sharp cutting edge. When a given cutting edge has dulled, it should at once be replaced by one or more fresh, sharp edges either through fracture or complete dislodgment of the dulled grit. If this replacement does not take place at or near the power level at which the cutting action takes place, because of too strong a bond or because of a loaded and hence reinforced condition in the wheel, then a rubbing action between grit

and metal becomes predominant (instead of merely transitory) and we may have considerable unwanted heat. When this rubbing action occurs, one may find that the continuous oil-phase type of grinding compound will maintain, through its lubricating value, a lower power level than cooler-operating aqueous types. Whether the high heat level at which these grinding oil systems operate does more or less damage to the metal than the "quenching" action of aqueous materials is a matter now being studied exhaustively. Oils are obviously impractical on most grinding operations because of this high heat level. But a nonloading grinding medium can be either aqueous or oily and specific advantages and disadvantages accrue to each method.

We are entering a new era in abrasives engineering. Indications are that the abrasive wheel as a precision tool will have its field of usefulness extended and more clearly defined through research and improved uniformity. The ceiling created by the inherent difficulty and anomaly of presenting sharp cutting edges concurrently with the dislodgment of dull grits at minimum power requirement will therefore be raised on many types of operations.

Written Discussion: By J. G. Morrison, chief metallurgist, Landis Machine Co., Waynesboro, Pa.

This paper might well be designated as required reading for all departmental heads having to do with grinding of hardened steels, particularly hardened tool steels. The losses due to faulty grinding are considerable and the dissemination of such information contained in Dr. Tarasov's paper should prove of great practical value.

One might gather the impression (unintended, I believe) from reading this paper that metallurgical factors loom fairly large as causes of grinding cracks. Our own experience in the case of hardened tool steels is that metallurgical factors are almost insignificant compared to the hazards surrounding the grinding operation itself.

The type of tool steel and an examination of the cleanliness, hardness and structure will generally suffice to gage the hazards due to metallurgical factors. However, so many details encompass the actual grinding, any one of which if incorrectly carried out may contribute to or cause grinding cracks, that, in the great majority of cases, the reason or reasons for grinding cracks have been found to be other than metallurgical. Many of the factors essential to correct grinding practice have been ably treated in the paper. One such factor, often overlooked and which does not appear to have been cited, is the skidding or tilting of the workpiece on a magnetic chuck.

On page 420 under "Other Factors" the greater susceptibility of high speed steel tempered at 500 degrees Fahr. (260 degrees Cent.) to check in grinding as compared to a 1050 degree Fahr. (565 degree Cent.) draw is believed due, primarily, to the instability of the residual austenite rather than to the presence of untempered martensite. A 500 degree Fahr. (260 degree Cent.) draw would temper the primary martensite to quite a considerable degree but leave untransformed not only most of the austenite, but that portion most sensitive thermally.

The author is to be complimented on the comprehensive treatment given the subject. Much can be said of its timeliness as economics will not permit the scrapping of expensive parts with the aplomb possible in war years.

Written Discussion: By Arnold L. Rustay, assistant chief metallurgist, Wyman-Gordon Co., Worcester, Mass.

Dr. Tarasov has summarized much of the published work on metal surface injury resulting from improper grinding and has taken the trouble to duplicate the work himself. There is plenty of evidence to confirm his statement that cracks and surface stresses in a ground surface are not necessarily the result of grinding practice inferior to that which normally produces a surface free from cracks, but there is a much greater weight of evidence that points to faulty grinding practice as the predominant cause of these defects.

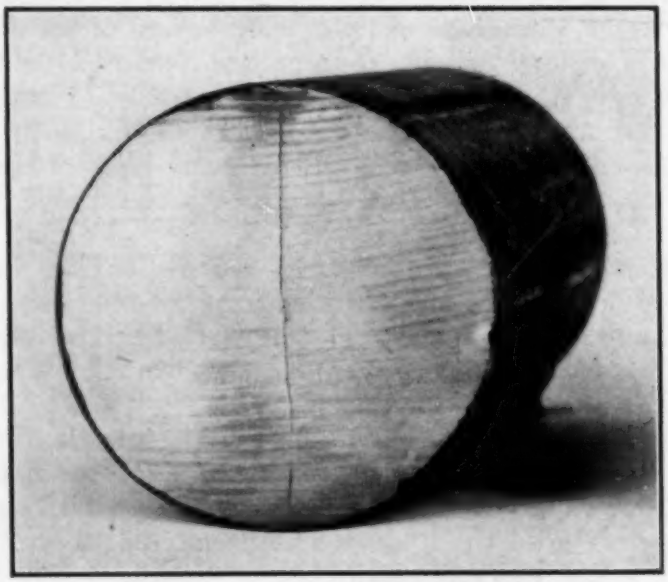

Fig. A—Showing Fine Crack.

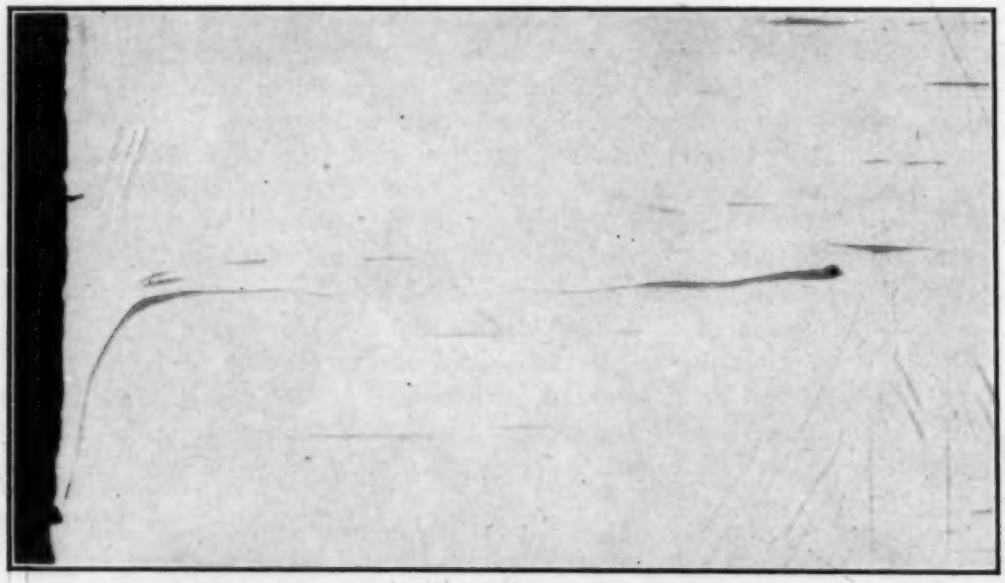

Fig. B—Section Normal to Crack. Distorted surface metal at the cut surface. Unetched. × 250.

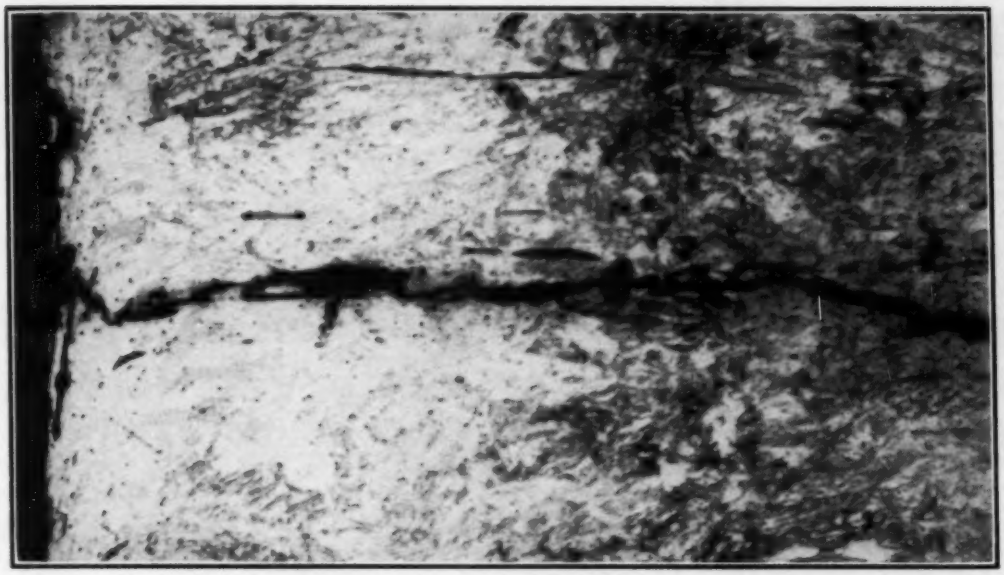

Fig. C—Typical Crack Extending from the Surface. Light area on left half is the band of hardened metal parallel to bar end. Etched. × 250.

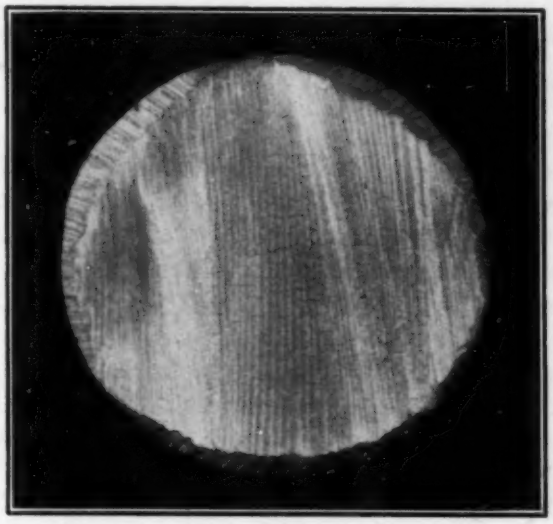

Fig. D—Fine Parallel Cracks at Right Angles to Cut-Off Wheel Marks. × 1.

The author discusses cracks and stresses in other hard metals, page 403, but does not mention the cracks that may occur in relatively soft metals when abrasive cut-off wheels are used. Hard, abrasive cut-off wheels, purposely chosen to give a large number of cuts per wheel, generate considerable heat at the cutting face even when using an ample supply of properly directed coolant. Hot-rolled bars of $2\frac{5}{16}$-inch round AMS-5610 (AISI 416) were cut to short lengths for forging. When the bar ends opened up during forging, examination showed cracks as shown in Fig. A. Fig. B shows the distorted surface metal at the cut surface and Fig. C shows a typical crack extending from the surface, and the band of hardened metal parallel to the bar end.

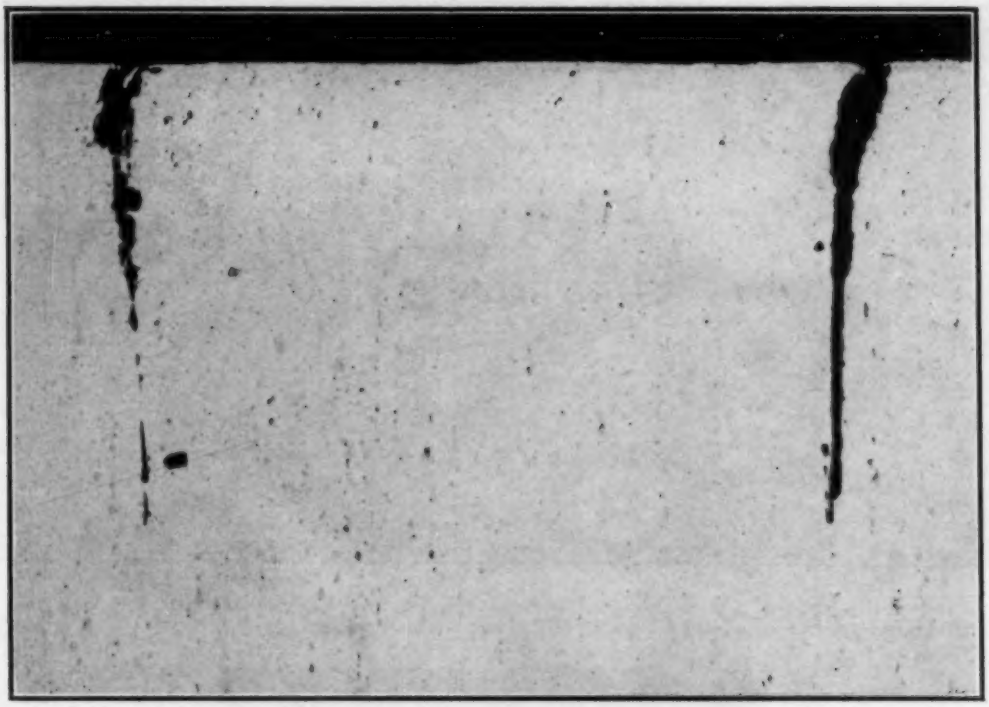

Fig. E—Same as D Showing Cracks Extending in from Surface. Etched × 100.

A less common cracking has been observed when a 75S aluminum bar has been cut in a similar manner. In this instance again, the cracks were found when the freshly cut bar ends were examined after the shop had had difficulty with end cracks during forging. Fig. D shows the fine parallel cracks at right angles to the cut-off wheel marks and Fig. E shows the cracks extending from the surface.

In both instances, the cracks were eliminated in subsequent lots by reducing the feed so that less heat was generated.

Oral Discussion

F. H. WELSH:[3] We have been working with grinding problems, and we have found out that it is quite possible to have grinding cracks using a very light cut on the finished product. Where you use a heavy cut then the grinding crack problem is almost eliminated. Can Dr. Tarasov tell me a a reason for that?

F. H. PENNELL:[4] I was very much interested in this paper. It confirms quite a number of things that I have experienced in the last few months in the grinding of reduction gear case-hardened worms. The retained austenite that was mentioned is proving to be a very undesirable factor, and it brings forth a controversial point that high hardnesses are more sensitive to grinding. I had some worms that, due to high retained austenite, were Rockwell C-50 to C-51, and they showed many more grinding cracks than ones that were thoroughly martensite and tempered back to about 60 or 61.

[3]Senior metallurgical engineer, Goodyear Aircraft Co., Akron, Ohio.
[4]Metallurgist, DeLaval Steam Turbine Co., Trenton, N. J.

There was one statement the author made about tempering of martensite before grinding at 300 degrees Fahr. I doubt if it is quite high enough. It might help considerably but I do not believe martensite begins to temper much under 325 degrees Fahr.

I certainly enjoyed the paper and wish to confirm a great many of the observations and findings.

PHILIP CONNOR:[5] Occasionally we had difficulty with the cracks on 430 grade stainless steel slab, 17 chromium, low carbon. These slabs are approximately 12 inches wide, 3 inches thick and about 30 feet long and the entire surface is ground cross-wise, and these cracks do not show up until after these slabs are reheated and rolled at least through two stands on a continuous mill, then they show up about 1 inch from the edge and extend in about 3 inches, but we have overcome this by giving it a light cut where we found heavy ridges on the ground surface. I would like to know if anyone has had an experience with this type of grinding cracks.

A. C. HARRIS:[6] We have been heat treating large quantities of small parts made of stainless steel, type 420. We experienced quite a bit of difficulty with grinding cracks in the parts after they were hardened and tempered and then finished ground. For a while the heat treating department was held responsible but after a representative from the Norton Company came into our plant the trouble was overcome. This was accomplished by changing various grinding wheels as the author suggests. However, I would like to know if the author has made any investigation into the grinding problems encountered in the 400 series stainless steels after hardening and tempering.

K. R. BLAKE:[7] There is one thing of importance to me, and I think to all of you other fellows, and that is the difference between stresses in the part and cracks or stresses on the surface. If we have a crack in a piece of steel, we are apt to find it by the Magnaflux. If we have tension stress in the surface, it is usually evidenced by distortion unless the cross section is great enough to offset it. But what about the stresses that are retained in the piece that are not shown by the Magnaflux? Those that are 5 per cent below the rupture strength of the metal and get by the Magnaflux offer the greatest possibility for a service failure. On the question of detection, I realize that there is no present day method for determining stresses retained in a piece of steel; that is, unless they evidence themselves by cracking or distortion, but we know they are present and I certainly think a lot of work should be done to find a method for determining the mechanism of those stresses and to overcome them, because certainly if a part gets out into service and is stressed very close to the rupture point and an overload is applied from any source whatsoever, a service failure will occur.

As to the effects of heat, Dr. Tarasov referred to heat generally and in the introduction to his paper he refers to heat and plastic deformation. Plastic deformation occurs in the cutting of a metal under certain conditions, but I do not believe it occurs to as great an extent with an increase in production rate. I would like to cite as an example the operation of negative rake milling. If you have had any experience with that, you know that a nega-

[5]Chief metallurgist, Farrell & Lowellville Works, Sharon Steel Co., Sharon, Pa.
[6]Supervisor of heat treat, Walter Kidde & Company, Inc., Belleville, N. J.
[7]Quaker Chemical Products Corp., Detroit.

tive rake mill which operates at many times the normal milling rate will machine a piece of steel with practically no evidence of heat. The heat and stress seem to vary inversely as the rate of production. I know that that is referring to a cutting tool that has a well defined cutting edge, but in the case of grinding I think the trouble is not in the plastic deformation, but rather is due to the fact that only a few of the grains are cutting and the balance of the grains are creating a rubbing action on the surface which results in the generation of heat. In the charts which Dr. Tarasov showed on the screen, you saw a wheel of H hardness, which showed no tendency to develop stresses whatsoever, and another of J hardness, in which the stress was quite high. In heat treatment, it is quite true that if you harden a piece of steel you get hardness by reason of introducing stress. You do not get a hard part without stress at the same time. Now certainly the harder the piece becomes, the greater the stress retained and the least safety factor we have left to permit the use of a grinding wheel which is noncutting. In other words, if we have a part that is Rockwell C-66, the part is bound to be stressed highly, because we could not have the hardness without the stress. In my opinion, the difference in the function of the wheel of H hardness and the wheel of J hardness is simply one of self-sharpening action. That is to say that the H wheel is soft enough to permit the grains to be released after only a minor degree of dulling, and thereby limiting the amount of rubbing action which would ordinarily have been created had the dull grains been retained. In the case of the J wheel, the amount of bond per grain has been increased sufficiently to prevent the dull grains from being released until a relatively higher pressure between the wheel and the work has been developed. The result, of course, is one of excessive metallurgical disturbance and heat production.

If we take a part which has been normalized and the austenite transformed to the very minimum, which I believe is around 3 to 5 per cent retained austenite, with sub-zero treatment, we will experience little or no difficulty in the grinding of this part without cracking. However, in my opinion, the part has been damaged due to the effect of the imparting of stress by the action of a dull grinding wheel. In other words, molecular changes have taken place, which have reduced the part's tolerance to overload. Furthermore, in reasonably severe cases of grinding of this type, a decided drop in the Rockwell hardness of the surface is usually found. This latter condition would certainly be objectionable in the case of a ball bearing race or similar part. Dr. Tarasov touched briefly on the question of taking heat away from a part which is being ground. He referred to the action of a grinding fluid as a "coolant". The term "coolant", in my opinion, is improperly termed since certainly if the grinding wheel is maintained in a free cutting condition so that surface rubbing action and wheel work pressure differential are kept to a minimum, large quantities of heat are not generated. Consequently, in any grinding operation where heat becomes a factor of material concern, I would expect to find the source of the trouble in the action of the grinding wheel itself.

This situation naturally brings up the question of what to do in the event excessive heat is produced. I would be interested in hearing Dr. Tarasov speak further on the question of wheel selection and application. However,

my guess is that a wheel that is free cutting enough to prevent the generation of heat will be very difficult to use in the maintenance of form. Therefore, we are faced with the compromise of trying to get the wheel to hold form and at the same time be free cutting. This question would bear serious consideration by the grinding wheel manufacturers. I might also include the consideration of the machine tool builders. Therefore, the harder the work being ground, the freer cutting the grinding wheel must be.

Author's Reply

It is most gratifying that the discussion has brought out additional information on injury in ground surfaces and thus made it readily available to everyone interested in this complicated subject.

In this connection, Mr. Doane's comments on the detection of surface cracks in hard metals other than steel are especially valuable since they are based on careful experimentation. The author's remarks on this phase of the subject were based to some extent on shop practice with which he was familiar but which can probably stand considerable improvement.

It is interesting to note that the fluorescent version of the magnetic-particle method has been applied to the detection of cracks in cemented carbides, even though these are only slightly magnetic. However, this technique can be used only if the carbide tip is not attached to the steel backing, and since most carbide tools are ground with the inserts brazed in place, it will not be possible, in general, to apply this method to detect surface cracks in ground carbide tools.

Mr. Moore's remarks bring out the importance of studying the action of grinding fluids with respect to surface injury. Experimental work now being carried out by some of the manufacturers of grinding fluids will undoubtedly lead to a better understanding of how these compounds act to reduce danger of injury to the ground surface.

Mr. Blake has raised the question of what can be done about stresses in ground surfaces. In addition to cracking and distortion methods for detecting such stresses, X-ray diffraction is a possibility if the steel is not too hard. In so far as determining the possible harmfulness of these stresses in service is concerned, fatigue testing is the obvious criterion. It should be kept in mind that even if harmful stresses have been introduced by severe grinding, they can be removed to a large extent by stress relief before the piece is placed in service. Burn, and the resulting soft skin, as in the example of the bearing race cited by Mr. Blake, cannot be cured once it occurs and hence must be avoided by grinding the part properly in the first place.

The statements about heat and plastic deformation, to which Mr. Blake has found an exception in the form of negative rake milling, were meant to be generalizations. It will generally be conceded that a heavy cut on a lathe develops more heat and deforms the metal to a greater depth than does a finishing cut. In negative rake milling the generalization does not hold because the cutting characteristics of the metal change as a result of the high speeds used.

Mr. Blake is correct in pointing out that the proper term to use is *grinding fluid* rather than *coolant,* which, however, has the sanction of common usage. The disadvantage of the latter term is that it implies to some people

that cooling is the sole function of the fluid whereas the compounds now generally used may be much more important from the standpoint of their oiliness characteristics. The question of correct wheel selection is obviously of the greatest practical importance but it is not possible to be specific about it since so many factors have to be considered in each grinding operation.

Mr. Morrison's experience has been that most grinding injury is caused by improper grinding practices and that metallurgical factors are relatively unimportant from a practical point of view. It is undoubtedly true in the majority of instances that grinding injury can be traced to a departure from correct grinding practice but there are exceptions. For example, we know of one large plant in the Detroit area in which grinding is so skillfully done and under such good control that only occasionally are ground surfaces found to be cracked; when this does occur, it is always traced back to improper heat treatment or to something the matter with the steel.

The reason that the author devoted what might be considered a disproportionate amount of attention to the metallurgical factors involved in grinding injury is that when these factors do happen to be responsible for the trouble, they are generally very difficult to detect and grinding gets the undeserved blame. Discussing the metallurgical factors in considerable detail should make is easier to do intelligent trouble shooting when such factors are involved, either primarily or partially.

As was pointed out by Mr. Morrison, the grinding sensitivity of high speed steel tempered at 500 degrees Fahr. (260 degrees Cent.) cannot be attributed to the presence of untempered martensite. What should have been said is that the sensitivity can be attributed either to the presence of retained austenite or to the formation of fresh martensite from this austenite during the grinding operation. As stated in the paper, we do not know at present which of these two phases is primarily responsible for the sensitivity.

Mr. Pennell's discussion is further proof that metallurgical factors may be very important in connection with the cracking of ground surfaces. In regard to tempering temperatures prior to grinding, the important thing is that the higher the temperature, the more severely the piece can be ground without cracking. Thus 350 degrees Fahr. (175 degrees Cent.) is better than 300 degrees Fahr. (150 degrees Cent.) for a plain carbon steel but if hardness requirements do not permit the higher temperature, the lower one will be perfectly satisfactory provided somewhat greater care is taken in grinding. The temperature at which martensite begins to temper, i.e., to transform from the tetragonal to the cubic form, and which is about 200 degrees Fahr., does not appear to be of any particular significance with respect to grinding sensitivity.

In reply to Mr. Welsh, if a light cut causes cracks to form while a heavier one does not, it indicates that the wheel is glazing (dulling) under the light cut and thus developing considerable frictional heat; the heavy cut, on the other hand, causes the wheel to break down and stay sharp. Possibly a softer wheel, which wears faster and presents fresh cutting points more frequently, is needed if the light cut is to be satisfactory.

Mr. Harris has asked about the grinding problems encountered in the hardened stainless steels, with special reference to surface cracks. We have not had the occasion to make such a study but we would expect the behavior

to be essentially the same as that described for other hardened and tempered steels. It is their martensitic condition that matters rather than their stainless composition.

The type 430 stainless steel mentioned by Mr. Connor is ferritic and not hardenable. We have never encountered in this type of steel any cracks that could be attributed directly to grinding. It appears from the description given that the heavy ridges left after grinding caused high stress concentrations to occur during the subsequent rolling, which were great enough to crack the surface. This is borne out by the fact that removal of the ridges by light grinding prevented the formation of these cracks. In this case, poor grinding practice would appear to be responsible for an undesirable surface contour which was directly to blame for the cracking.

Mr. Rustay's two examples have to do with cracks in soft materials caused by the abrasive cut-off operation, which is an extremely severe form of grinding. The type 416 stainless steel was heated high enough during cutoff to deform plastically and to transform to austenite so that it hardened to a considerable depth on cooling. The cracking occurred in the hardened layer so that the behavior can be considered to be essentially like that observed in hardened steels ground too severely. The aluminum bar, on the other hand, remained soft throughout the whole cutting-off operation. The appearance of these cracks in an aluminum alloy is of great interest as being the first published instance of harmful cracks in a soft metal subjected to some form of abrasive operation. Wulff (15) had previously shown the presence of barely detectable microscopic cracks in the ground surface of beta bronze, but these were too tiny to be of any practical importance.

ANTI-REFLECTION FILMS FOR METALLOGRAPHIC OBJECTIVES

By James R. Benford

Abstract

Experimental studies on improvement in performance of metallographic microscopes due to anti-reflection films on the objective lens surfaces show the improvement to be dependent on the objective design and on the type of specimen viewed. The improvements accomplished by the filming consist of a gain in image contrast and a shortening of the photographic exposure. Photomicrographs are submitted showing comparative performance between filmed and unfilmed objectives. These results are supplemented by observations made with a visual comparator device which enables the observer to view two metallographic microscope images simultaneously. Photoelectric measurements of percentage flares in the images are correlated with the photographic and visual results.

THE application of anti-reflection films to objective lens surfaces decreases reflected flare and increases the light transmission. For an objective used in vertical illumination this means that, if filmed for anti-reflection, it will transmit more light to the specimen and on the return passage will transmit more light back to the image. At the same time the unwanted reflected flares from the lens surfaces will be reduced. A gain in contrast is to be expected. This paper describes experimental evaluations of this gain in contrast.

To obtain a direct comparison between filmed and unfilmed objectives, eleven matched pairs of objectives of different powers were selected from stock. One member of each pair was disassembled and filmed for anti-reflection, after which it was reassembled and checked to determine that its image quality was still the equal of the unfilmed member of the pair. In order to divorce contrast from this test (as far as possible) the comparison was made in transmitted light instead of in vertical illumination.

The filming was done by the high vacuum thermal evaporation

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process. For this work a type of anti-reflection film was selected which, while not as efficient as some, is of exceptional hardness. This point must be borne in mind in considering the results reported in this paper, since the hard type of film reduces the flare by a factor of about 3 to 1 only, whereas it is easily possible to reduce flare by a factor of 4 or 5 to 1 by the use of softer films. Practical considerations in regard to handling, working, and cleaning the exceptionally tiny lenses used in microscope objectives led to the choice of the harder films, despite the lower efficiency. Future trends toward films which are both hard and of high efficiency, or toward improved methods of handling filmed objectives, will undoubtedly result in the commercial availability of filmed objectives giving better performance than the ones reported in this present work.

Since metallographic objectives are used for both visual and photographic work, comparisons were made both visually and photographically. These results were qualitative in nature and depended on the observer's discrimination as to "how much" improvement existed in the visual or photographic image due to the anti-reflection films on the objective. Complementing this work a photoelectric comparison was made, giving numerical results on flare evaluation of a purely objective nature.

All results reported in this paper were obtained using centered vertical illumination. A plane glass vertical illuminator was used. Bright field illumination results obtained with the Foster[1] calcite vertical illuminator prism should be very closely comparable to results with the plane glass vertical illuminator. In dark field or polarized light the flares are much less than in bright field.

Visual Comparison

The set-up used for making the visual comparison is shown in Fig. 1. This system utilized a comparison eyepiece to present the observer with two images side by side for direct comparison. The two sides of the system were made as nearly equal as possible, with the exception that one side used the filmed objective and the other the unfilmed objective.

To get the two specimens as nearly alike as possible a single specimen was ground, polished, and etched, after which it was cut in two vertically, one half being used on one side and the remaining

[1]L. V. Foster, *Journal*, Optical Society of America, Vol. 28, April 1938, p. 124-129.

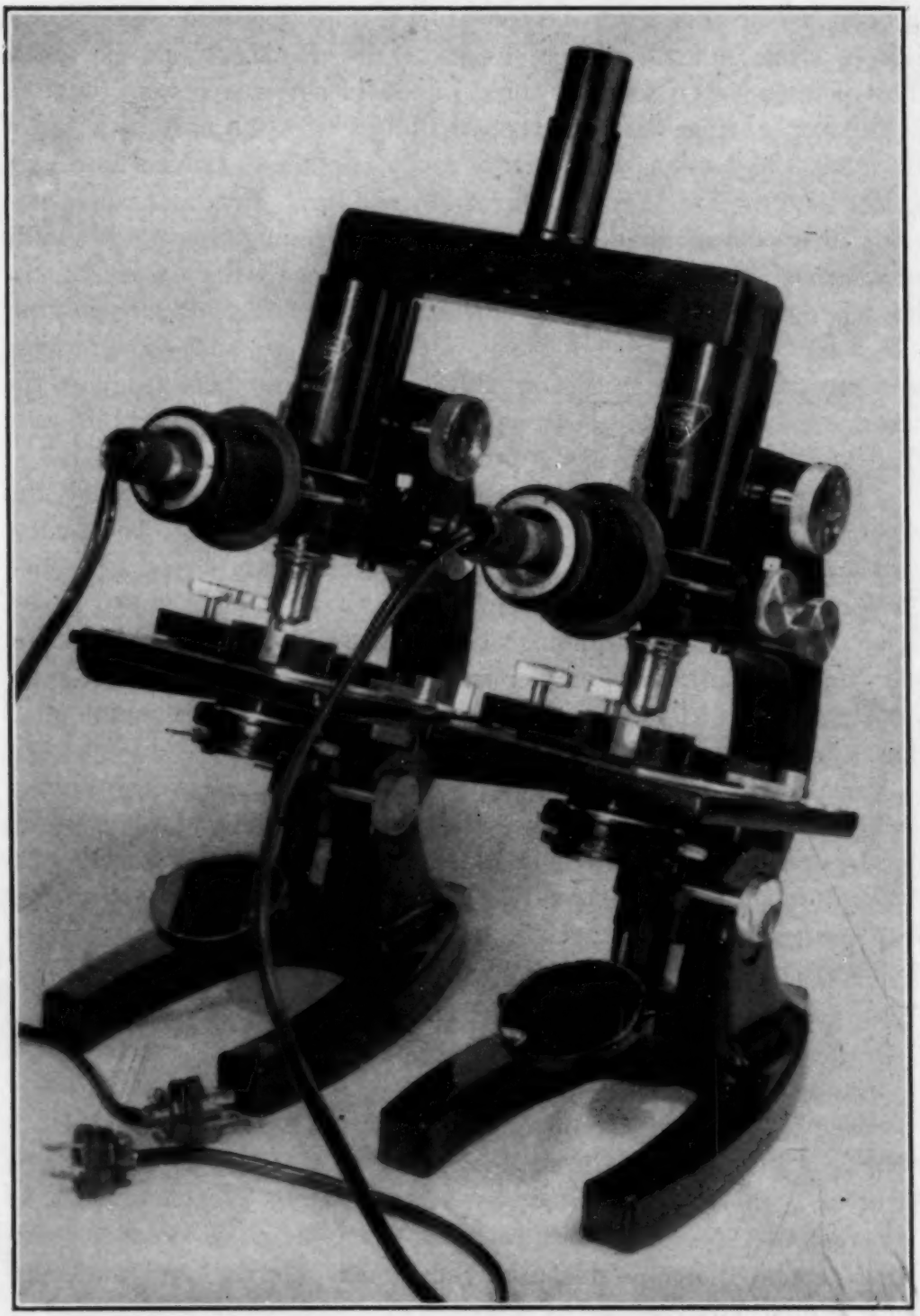


Fig. 1A—Apparatus for Visual Evaluation. The two systems are identical except that one uses a filmed and the other an unfilmed objective. The two images are seen in juxtaposition in the eyepiece field as shown in Fig. 1B. A single finished specimen is cut vertically to provide two closely similar specimens.

half on the other. To make the test more conclusive, the objectives were mounted in quick-change mounts so that they could be easily switched from one side to the other.

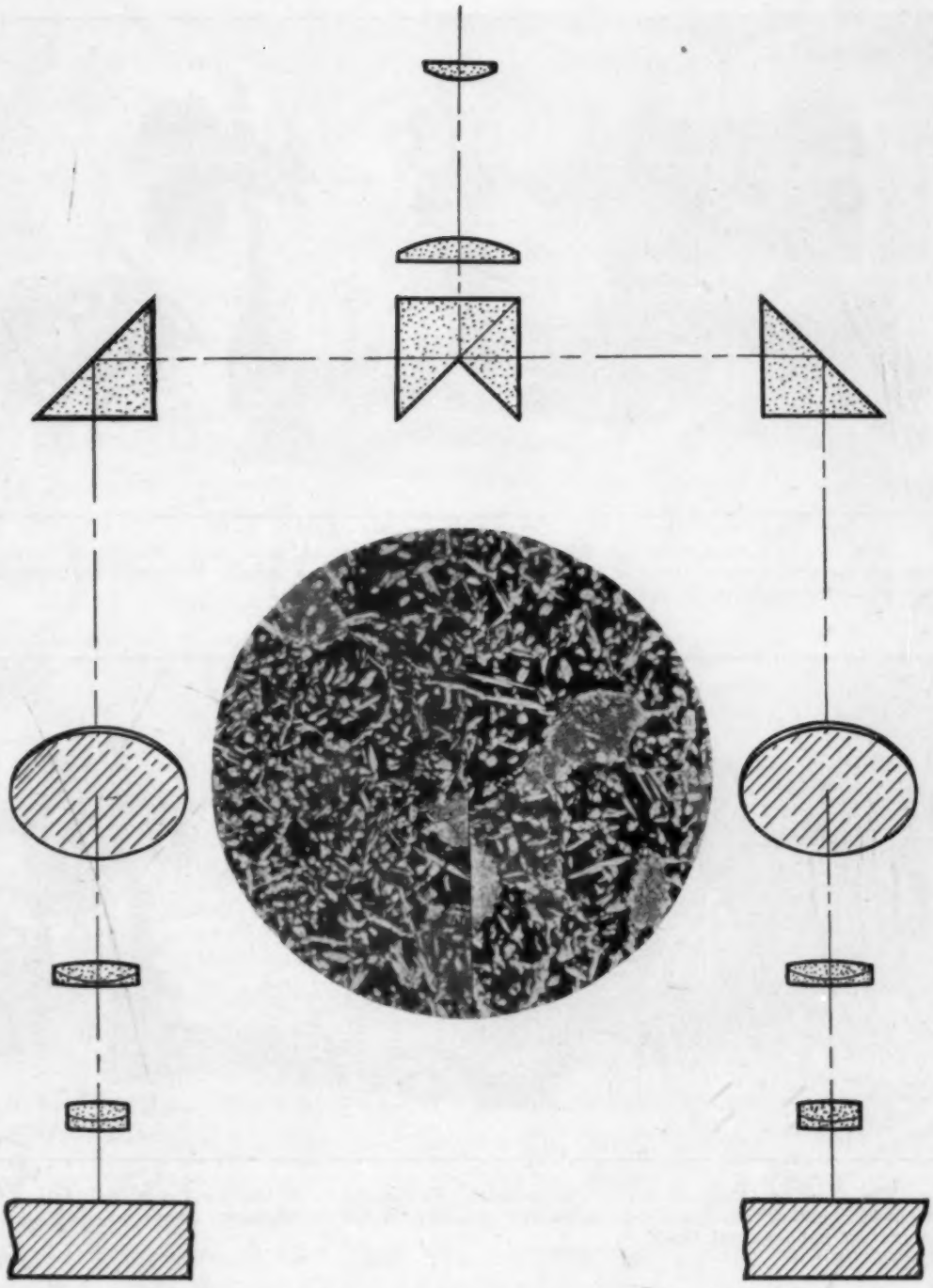


Fig. 1B—Schematic Drawing of the Optical System of the Visual Comparator. Insert shows the split field of view obtained with this instrument. Specimens are manganese bronze in this case.

Photographic Comparison

To compare the photographic results between filmed and unfilmed objectives the small metallographic microscope shown in Fig.

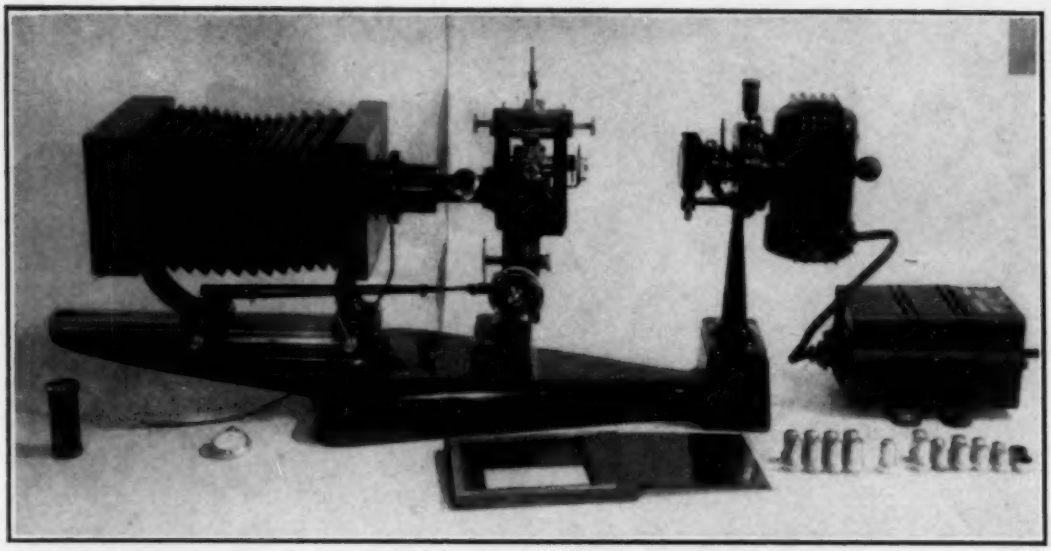


Fig. 2—The Apparatus for Photographic Evaluation. A split mask interposed between the photographic plate and the regular dark slide provides a means for obtaining comparison photomicrographs on the same negative.

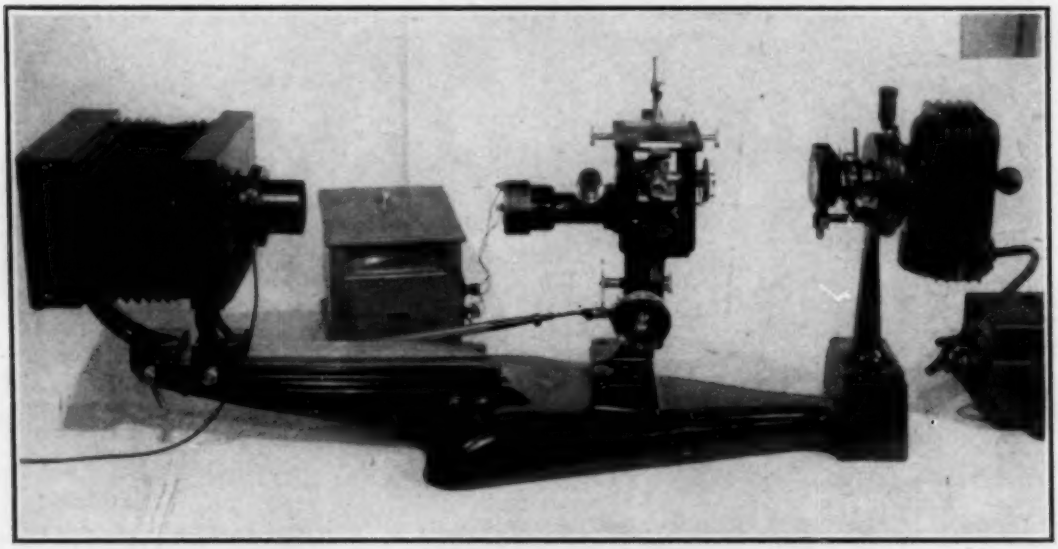


Fig. 3—The Apparatus for Photoelectric Evaluation. A barrier layer cell and 0.01 μA/mm. galvanometer provide sufficient sensitivity for measuring objective flare as a percentage of the total light.

2 was used. This was equipped with a special plate holder as shown, having, in addition to the normal dark-slide, a second dark-slide cut away to cover only half the plate. This latter slide was used to mask half the plate for a photomicrograph with the unfilmed objective, and then flipped over to mask the other half of the plate for a photomicrograph with the filmed objective. The specimen was shifted in between these exposures, so that the two pictures showed the identi-

cal area in each case. This device assured the same photographic technique for both pictures, since both were developed on the same negative and printed simultaneously on the same paper.

The two pictures had to be closely equal in exposure. The photoelectric device shown in Fig. 3 was used as an exposure meter to get the relative exposures required for filmed and unfilmed objectives, and the lamp was operated on a constant voltage transformer.

The photographic technique selected as being most conclusive was to aim for a print which neither exaggerated nor subdued the image contrast. In other words, the product "gamma" of negative and print was to be approximately unity. To accomplish this W. & W. metallographic plates were tray developed in D41 for 4½ minutes and contact prints were made on Grade 0 paper. According to the manufacturer's data both negative and print should be close to a gamma of unity under the above conditions.

All photomicrographs were taken with the yellow-green filter in the illuminating beam.

Photoelectric Comparison

The photoelectric apparatus used is shown in Fig. 3. The photocell was an "Electrocell" and the meter a 0.01 μA/mm. Rubicon galvanometer. A simple ring adapter was used to cup the cell onto the light-tight connector surrounding the camera eyepiece.

In evaluating the flare, the procedure adopted was to measure the ratio of flare, or unwanted light, to the total light reaching the image plane. The flare was then expressed as a percentage of the total light.

To measure flare alone, a zero reflection specimen was needed. Black velvet held an inch or so above the objective served very well for the dry objectives. For the immersion objectives, a disk of glass, polished on the lower surface and ground and blackened on the upper surface, was used. This was oil-contacted to the objective in measuring the flare.

In measuring the total light going into the image, the specimen was brought into focus in the visual eyepiece and this eyepiece then withdrawn, permitting the light to strike the photocell. The 10X negative amplifier was used in the camera eyepiece tube since this gave a good large spot of light on the photocell, and made the reading less sensitive to slight position changes in the cell. The yellow-green

filter was not used for the photoelectric readings as it reduced the sensitivity too much.

Evaluation of the Results

Each of the three methods, visual, photographic, and photoelectric, has its advantages and its limitations. The photoelectric method is the only one which gives quantitative results, but it takes no account of the contrast range in the specimen, merely giving an average value for the whole area of the specimen in the field of view. The visual method requires two specimens which can never be wholly identical. Furthermore the results are convincing only to the man looking into the instrument, and word descriptions carry little weight in a publication. The photographic method yields results capable of reproduction in publication form, but as no photographic reproduction is perfect, some loss of tone range is inevitable in the finished print. This is accompanied by a further loss in the half-tone process. For this reason the improvement seen by the visual set-up is greater than can be reproduced photographically.

It will be appreciated that the results depend to a great extent on the type of specimen chosen. The flare coming from any one of the objectives is a constant, but the useful image-forming light coming from the specimen depends on the reflection coefficient and surface etch condition of the specimen. Flare is most obvious with dark and deeply etched specimens, and by the same token, reduction of the flare by filming is most easily seen with such specimens.

The inherent contrast in the specimen must also be considered in evaluating the results of filming. The effect of objective flare is to reduce contrast since it adds a more or less even haze over the whole picture, and at the same time reduces the useful light going to the specimen. The image of a specimen which has only extreme contrasts will not appear to be greatly improved by the filming, while that of a specimen having many soft shades of gray will be noticeably improved, particularly if it contains dark shades which must be differentiated. The loss of contrast in the darker regions is more noticeable because there the ratio of flare to useful light is greater. The presence of high lights as well as low lights in the picture also makes the reduction of flare more noticeable because there is then more light contributing to the flare in the low lights, and also a greater tone range to be captured in the photographic image. This latter fact means that photographic "dodges" to overcome the flare are

rendered difficult or sometimes impossible to properly manipulate.

Figs. 4, 5, 6, and 7 demonstrate the preceding argument as to the pronounced effect of the specimen nature on the comparative improvement due to filming. Each of these photomicrographs was taken with the same pair of 32-mm. objectives, yet the improvement due to filming is widely different in the four pictures. The aluminum specimen, Fig. 4, has very few shades of gray, it is contrasty, and has a high reflectivity (about 71 per cent). The improvement due to filming is rather slight. The annealed alpha brass, Fig. 5, is contrasty but has a lower reflectivity (about 38 per cent), and shows more improvement than the aluminum. The manganese bronze, Fig. 6, is a contrasty specimen, but its reflectivity is still lower (about 17 per cent) and the improvement due to filming is very marked. The fourth specimen, chromium plating for a bearing surface,[2] Fig. 7, is an extreme type, having a reflectivity of only about 2.5 per cent. The flare is extremely pronounced in the unfilmed shot, and the gain in contrast due to filming is quite striking.

Thus the decision as to the advisability of filming the objectives hinges not only on the objective designs but also on the types of specimen encountered in metallographic work. If the metallurgist frequently, or even occasionally, encounters specimens of low reflectivity and long tone range, filming would appear advisable. On the other hand if almost all of the metallurgist's work is on specimens of high reflectivity and sharp contrasts, filming will be of small advantage.

Cast iron has been found to be a useful specimen to show the comparative performances of various objectives. It has detail at practically any magnification, and has a good tone range. Its average reflection coefficient is quite low—about 7 per cent—so that flare is easily seen if present. It should be emphasized that cast iron is undoubtedly not a typical specimen because its reflectivity is low for a metal specimen. As comparative figures, the following reflectivities were measured on individual metal samples:

	Per Cent
Fine untempered martensite in quenched drill rod	10
Pearlite	16
Spheroidized cementite	28
Cold-worked brass	28.5
Coarse martensite in partially transformed air-hard steel	36
Aluminum	71

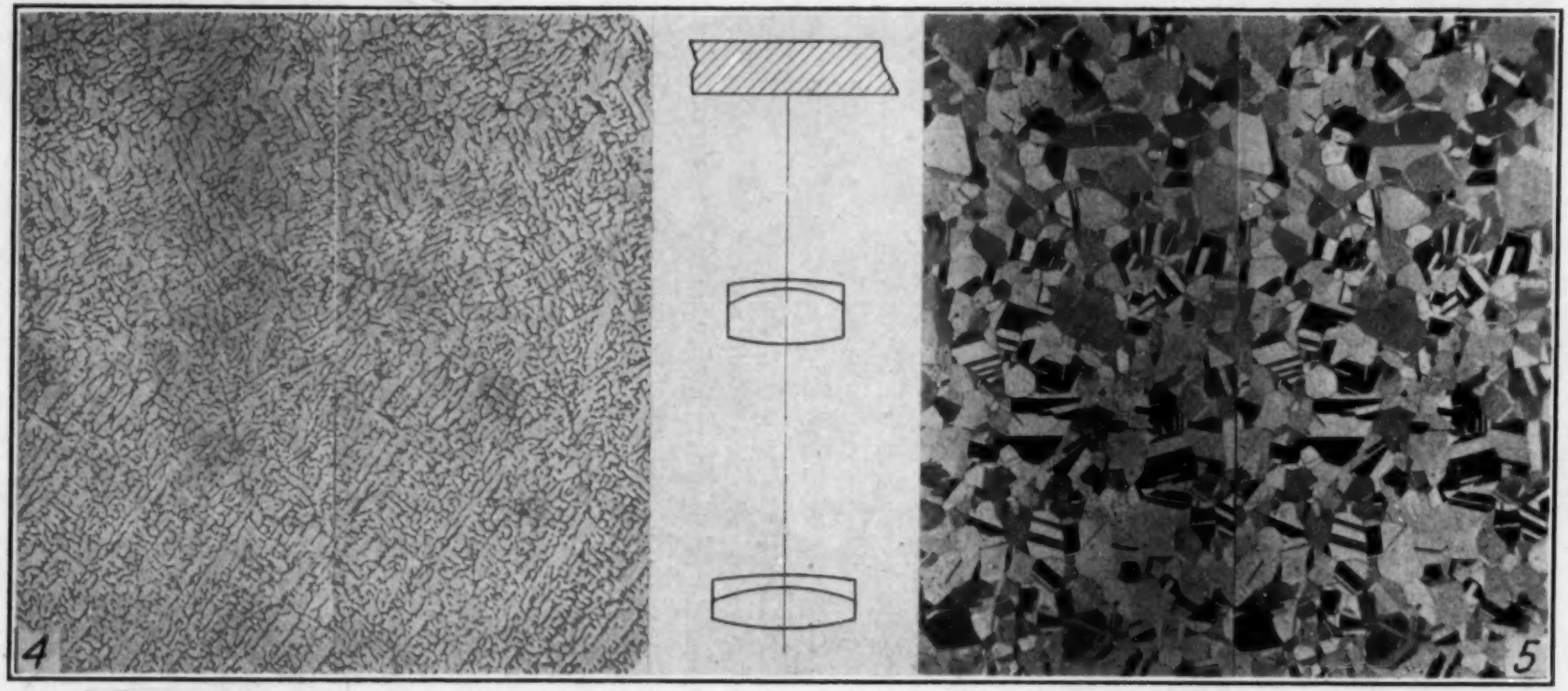
These values are merely results on individual specimens, and are intended to give a rough indication of their comparison with cast iron.

[2]This specimen of porous chromium was obtained through the courtesy of Mr. A. H. Baum, Van der Horst Co., Olean, N. Y.

Cast Aluminum, × 50

32 mm., 0.12 N.A.
Full Aperture

Annealed Alpha Brass, × 50



Unfilmed
Flare 7.9%

Filmed
Flare 2.3%

Specimen Reflectivity 71%

Fig. 4

Objective Design

Unfilmed
Flare 12.5%

Filmed
Flare 3.4%

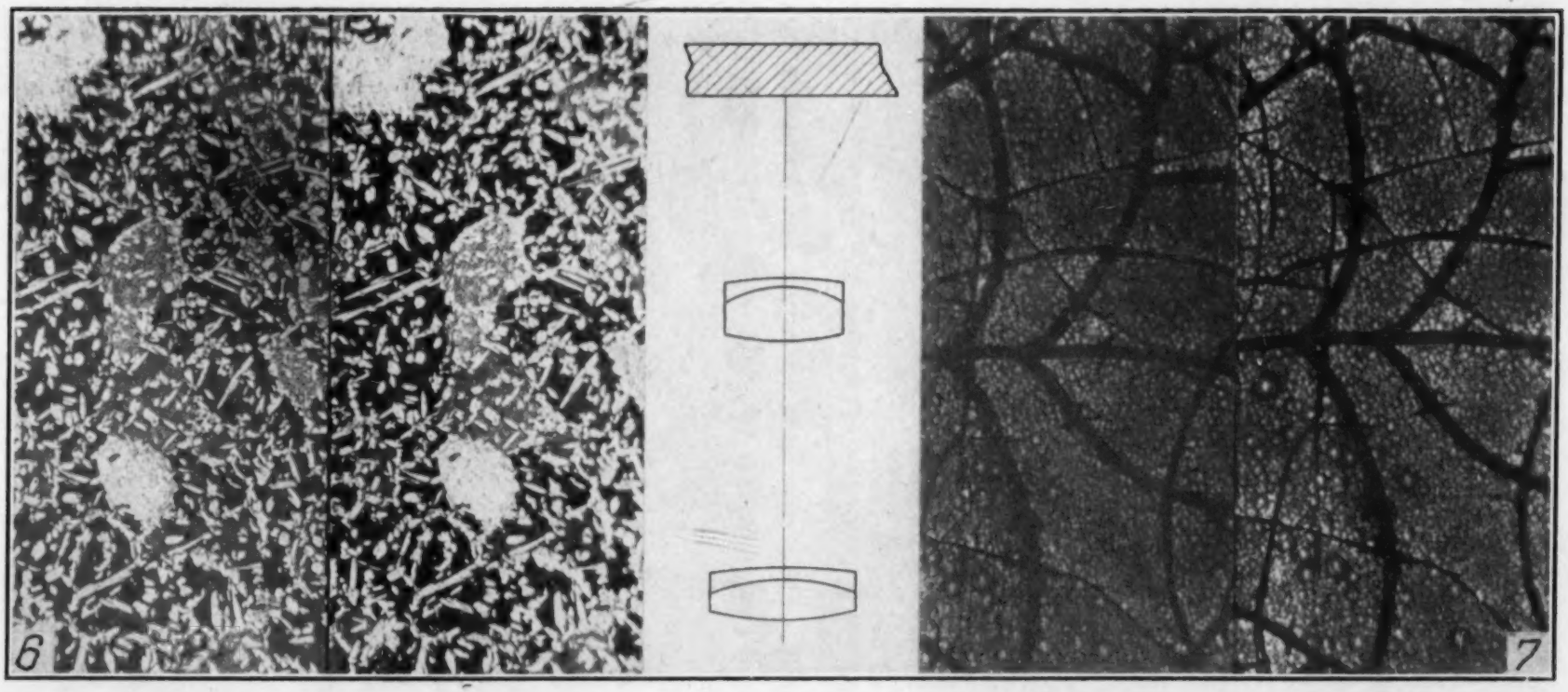
Specimen Reflectivity 38%

Fig. 5

Manganese Bronze, × 50

32 mm., 0.12 N.A.
Full Aperture

Chrome Plate for Bearing Surface, × 50



Unfilmed
Flare 29%

Filmed
Flare 9%

Specimen Reflectivity 17%

Fig. 6

Objective Design

Unfilmed
Flare 66.5%

Filmed
Flare 32%

Specimen Reflectivity 2.5%

Fig. 7

Figs. 8 to 18 show the identical region—at varying magnifications—in a cast iron specimen photographed with each of eleven pairs of different power objectives. The photographic evaluation which appears in Table I was based on the photographic print—not the half-tone—and represents one individual's opinion only.

The visual evaluation in Table I is based on results obtained with the apparatus shown in Fig. 1, and again represents one individual's

Table I
Data for Photomicrographs of Cast Iron, Figs. 8-18

Fig. No.	Mag.	Objective	Aperture	Photoelectric Flare Values Left Field Unfilmed (Per Cent)	Right Field Filmed (Per Cent)	Visual Improvement	Photographic Improvement
8	50	32 mm., 0.10 N.A., Achromat	Full	4.3	1.1	Definite	Slight
9	50	32 mm., 0.12 N.A., Achromat	Full	42.5	13.6	Extreme	Extreme
10	100	24 mm., 0.20 N.A., Achromat	Full	20.7	5.9	Definite	Definite
11	150	16 mm., 0.25 N.A., Achromat	Full	17.5	4	Definite	Definite
12	250	10.25 mm., 0.40 N.A., Achromat	Full	12.0	3	Definite	Definite
13	500	6 mm., 0.65 N.A., Achromat	Full	4	0.8	Slight	Negligible
14	750	4 mm., 0.95 N.A., Apochromat	4/5	20	4.9	Definite	Definite
15	1000	4.3 mm., 1.00 N.A., Fluorite	Full	9.1	2.7	Slight	Negligible
16	1500	1.8 mm., 1.25 N.A., Achromat	4/5	24.5	8.2	Definite	Definite
17	1500	2.75 mm., 1.25 N.A., Fluorite	4/5	4.2	1.8	Slight	Negligible
18	1500	3 mm., 1.40 N.A., Apochromat	2/3	13.4	4.6	Definite	Definite

opinion only. The cast iron specimen photographed in the series, Figs. 8 to 18, was used on one side of the visual comparator, Fig. 1, and the other half of the same specimen was used on the other side of the comparator.

The photoelectric evaluation of the flare appears as a percentage value below each picture. *These flare values refer only to the relative flare for the cast iron specimen,* e. g., in Fig. 13, the flare value of 4 per cent means that the flare coming from the objective lenses alone is 4 per cent of the total light coming from both the objective and the cast iron specimen. With a specimen of higher reflectivity the flare values would of course be less. From a comparison of the photoelectric values and the visual and photographic evaluations, it would appear that if the unfilmed member of the pair shows a flare value less than about 5 per cent of the total light, the photographic gain from filming will be negligible, but some visual gain will still be achieved. If, on any specimen, the flare value of the objective goes as high as about 12 per cent, photographic improvement appears to be definite.

The cause of flare in a lens design would appear to be traceable to: (1) the number of air-glass surfaces, and (2) the concentration

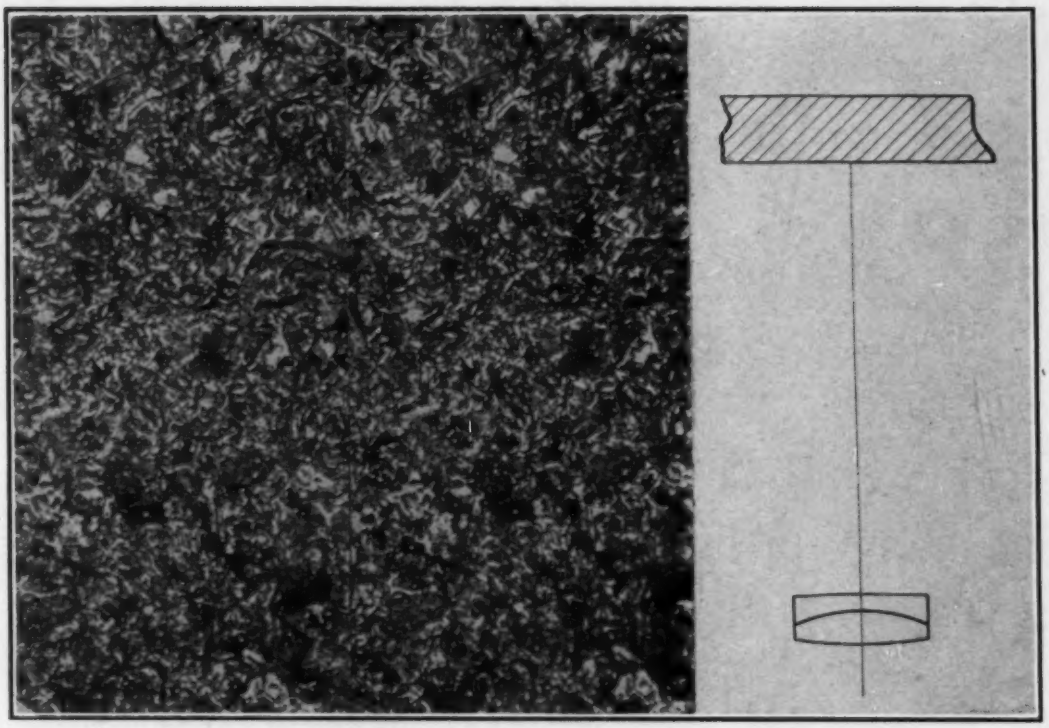


Fig. 8—Photomicrographs of Cast Iron. × 50. Showing objective design. 32 mm., 0.10 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 4.3%. Right field, filmed objective, flare = 1.1%.

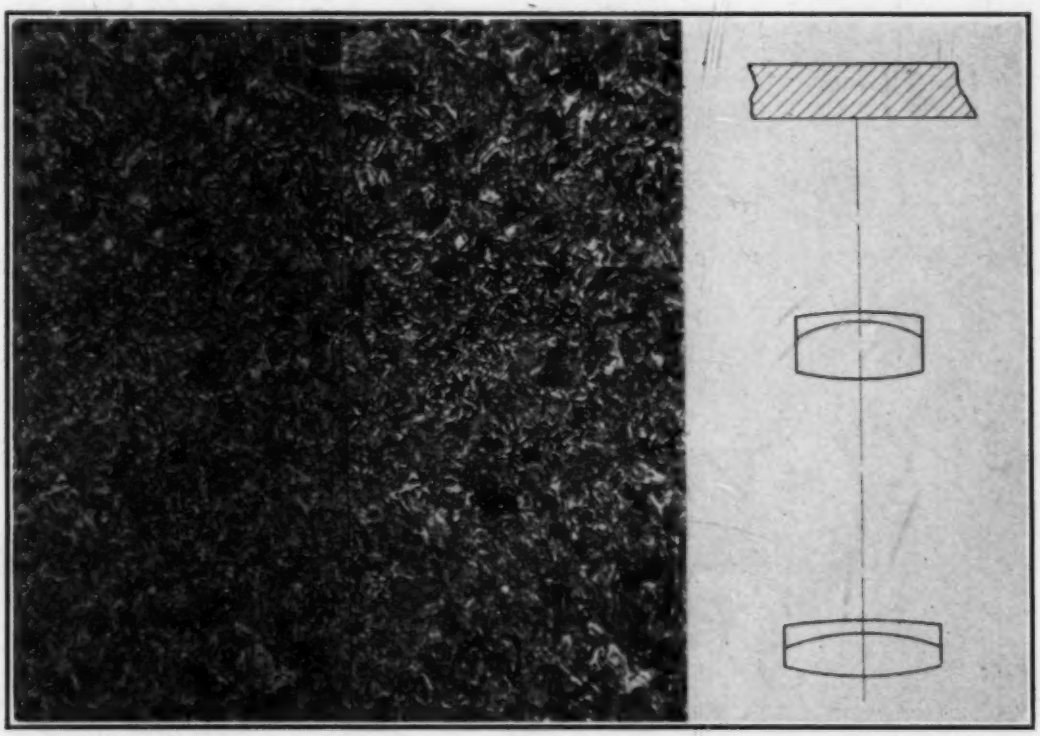


Fig. 9—Photomicrographs of Cast Iron. × 50. Showing objective design. 32 mm., 0.12 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 42.5%. Right field, filmed objective, flare = 13.6%.

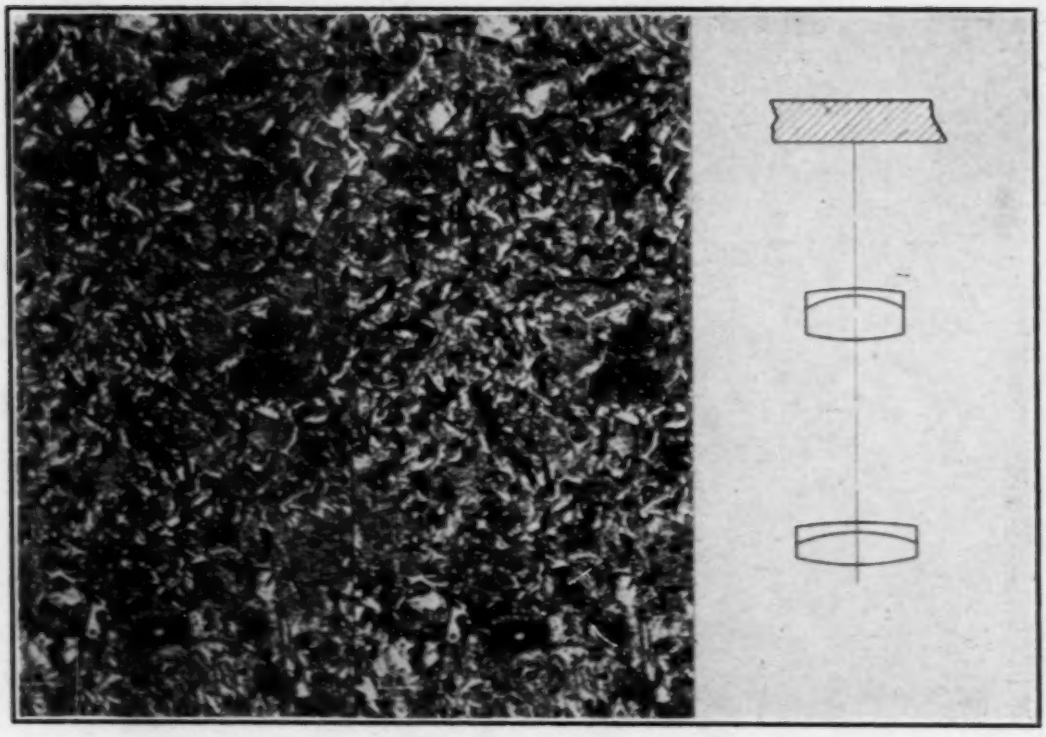


Fig. 10—Photomicrographs of Cast Iron. × 100. Showing objective design. 24 mm., 0.20 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 20.7%. Right field, filmed objective, flare = 5.9%.

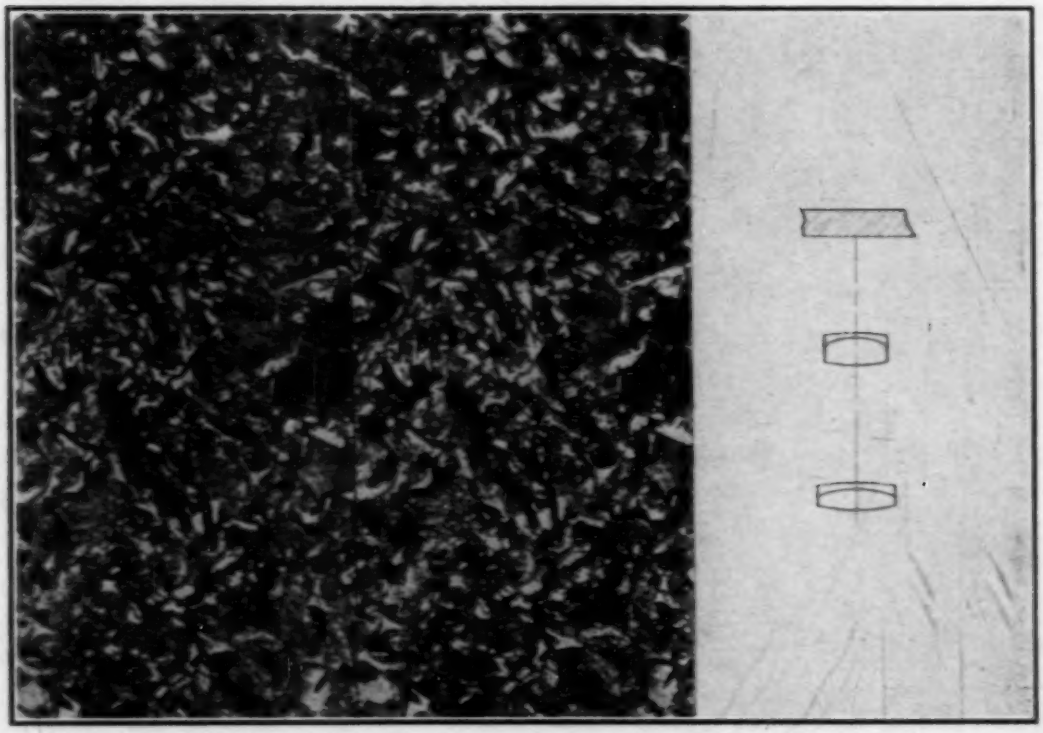


Fig. 11—Photomicrographs of Cast Iron. × 150. Showing objective design. 16 mm., 0.25 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 17.5%. Right field, filmed objective, flare = 4.0%.

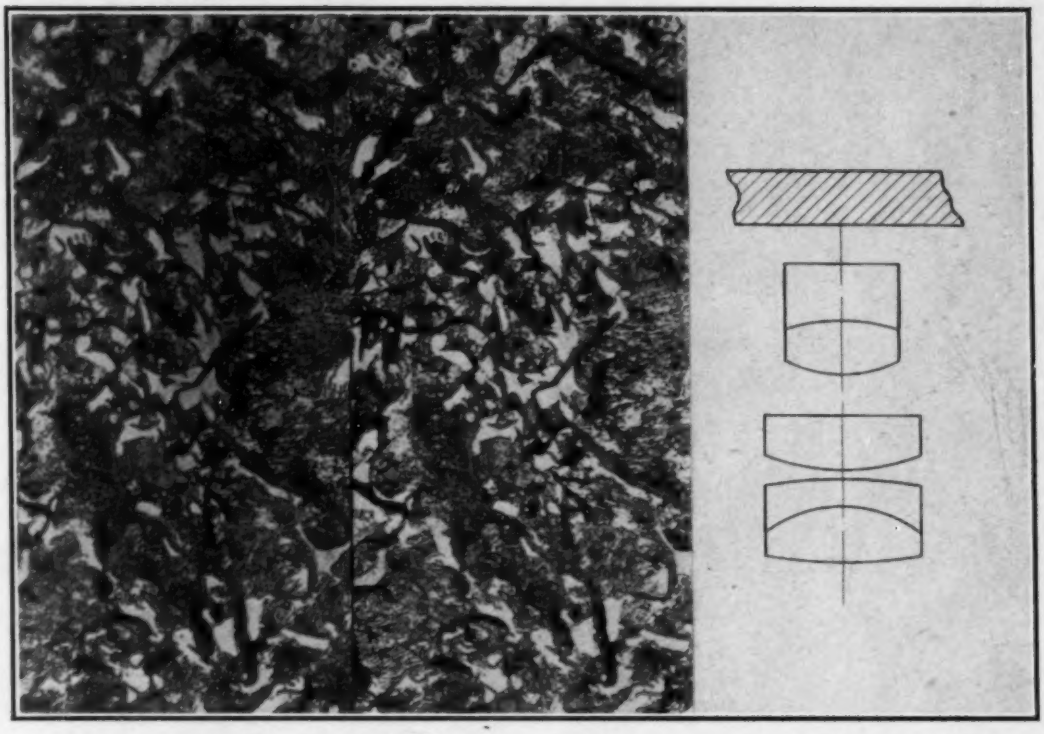


Fig. 12—Photomicrographs of Cast Iron. × 250. Showing objective design. 10.25 mm., 0.40 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 12.0%. Right field, filmed objective, flare = 3.0%.

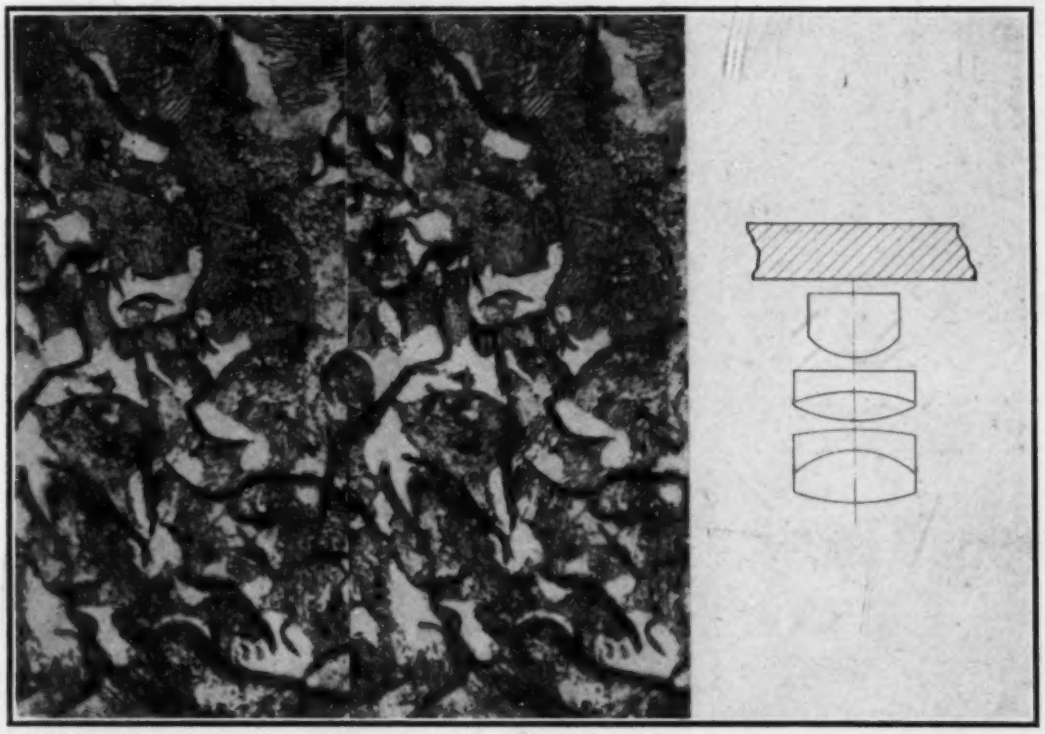


Fig. 13—Photomicrographs of Cast Iron. × 500. Showing objective design. 6 mm., 0.65 N.A., Achromat. Full aperture. Left field, unfilmed objective, flare = 4.0%. Right field, filmed objective, flare = 0.8%.

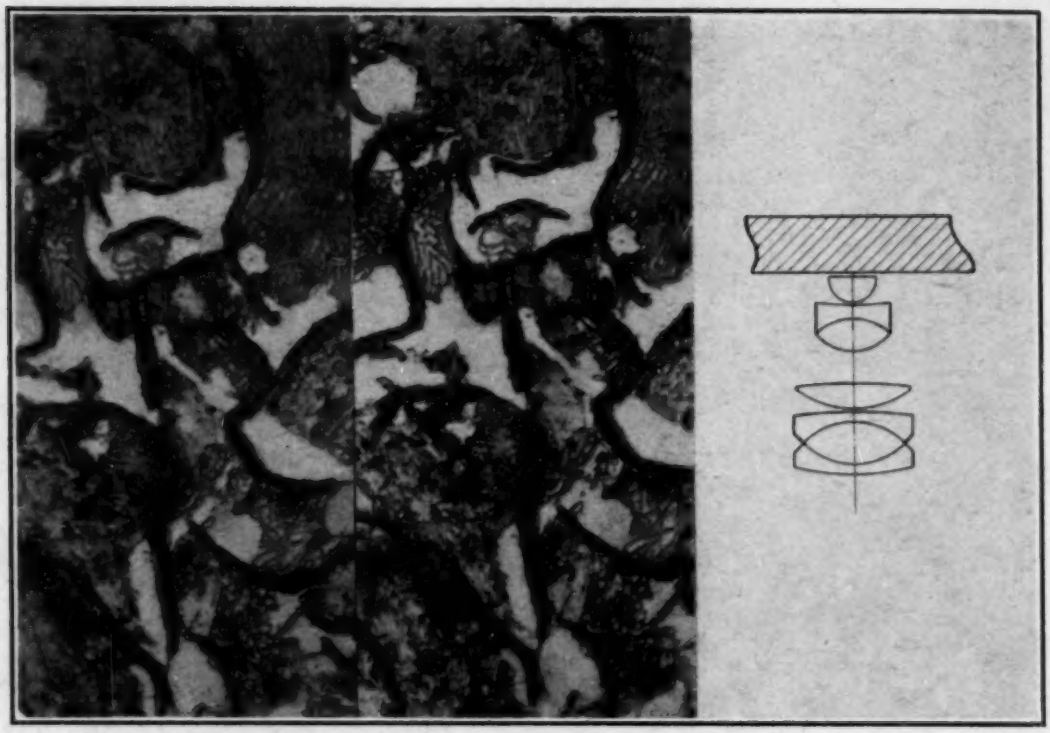


Fig. 14—Photomicrographs of Cast Iron. × 750. Showing objective design. 4 mm., 0.95 N.A., Apochromat. ⅘ aperture. Left field, unfilmed objective, flare = 20.0%. Right field, filmed objective, flare = 4.9%.

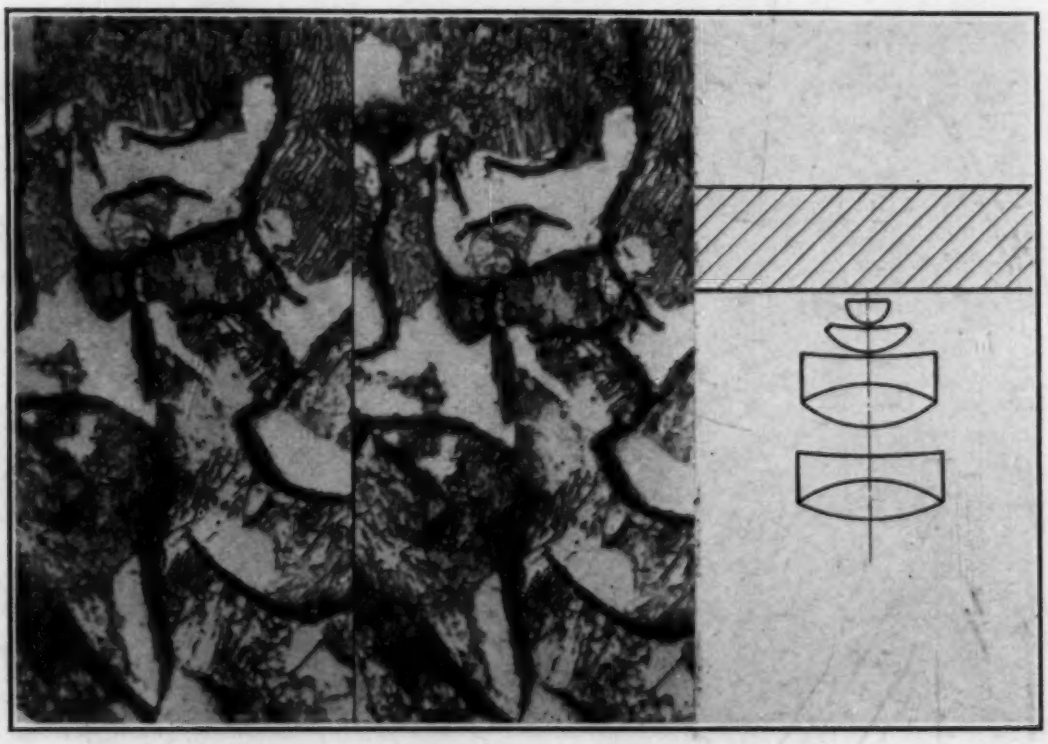


Fig. 15—Photomicrographs of Cast Iron. × 1000. Showing objective design. 4.3 mm., 1.00 N.A., Fluorite. Full aperture. Left field, unfilmed objective, flare = 9.1%. Right field, filmed objective, flare = 2.7%.

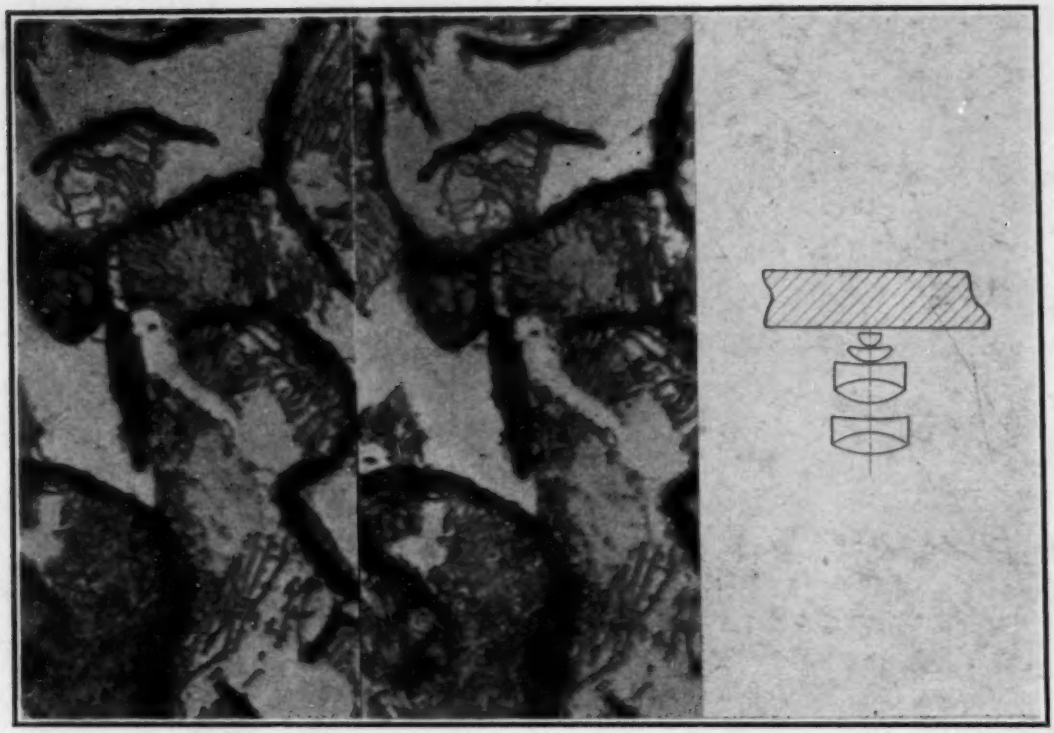


Fig. 16—Photomicrograph of Cast Iron. × 1500. Showing objective design. 1.8 mm., 1.25 N.A., Achromat. ⅘ aperture. Left picture, unfilmed objective, flare = 24.5%. Right picture, filmed objective, flare = 8.2%.

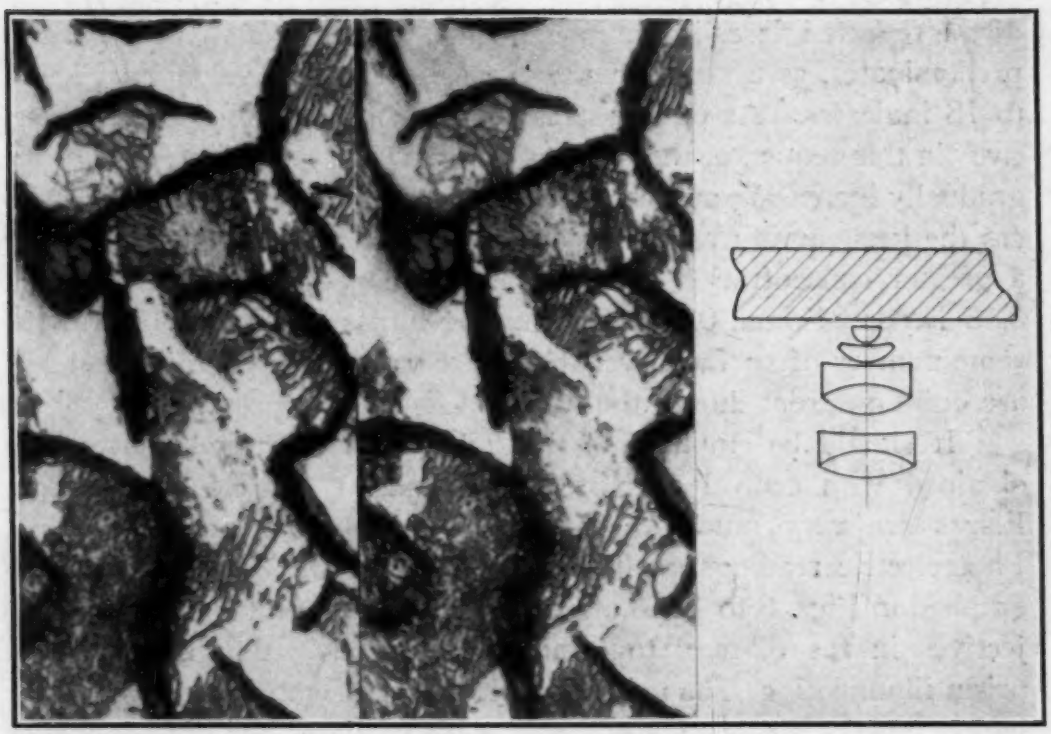


Fig. 17—Photomicrographs of Cast Iron. × 1500. Showing objective design. 2.75 mm., 1.25 N.A., Fluorite. ⅘ aperture. Left picture, unfilmed objective, flare = 4.2%. Right picture, filmed objective, flare = 1.8%.

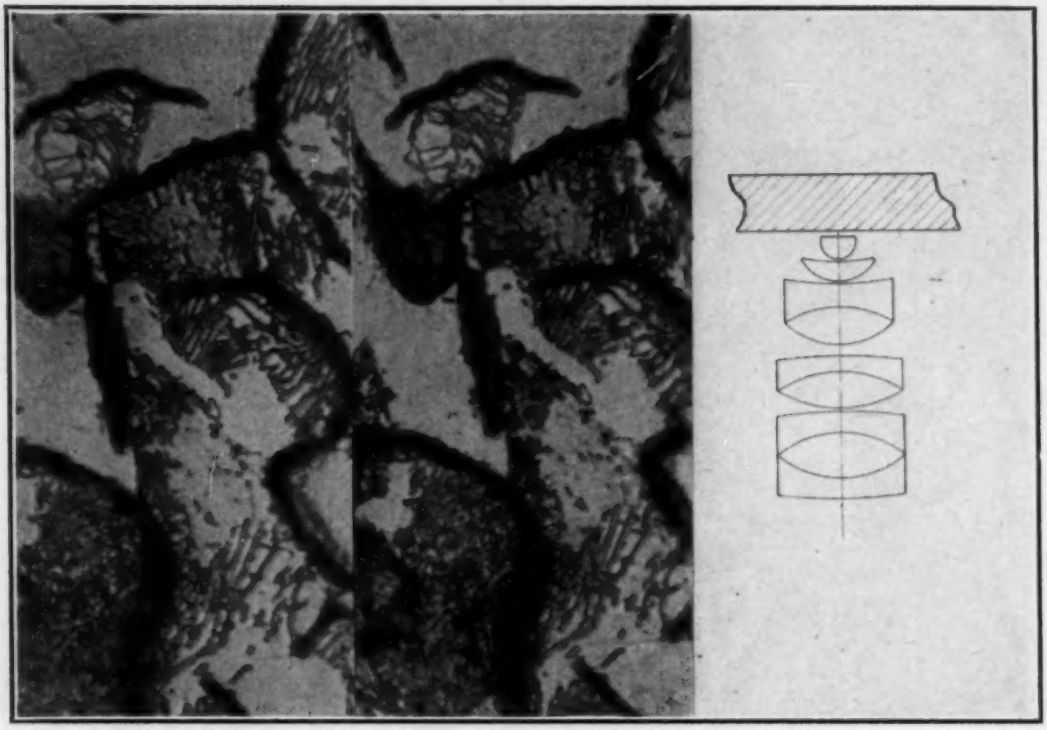


Fig. 18—Photomicrographs of Cast Iron. × 1500. Showing objective design. 3 mm., 1.40 N.A., Apochromat. ⅔ aperture. Left field, unfilmed objective, flare = 13.4%. Right field, filmed objective, flare = 4.6%.

of the reflected flares in the image plane. The latter effect in general predominates, as a study of the photoelectric flare values in Figs. 8 to 18 indicates. As one progresses from low to high power objectives in this sequence, the flare *trend* is toward reduction, despite the gradually increasing number of air-glass surfaces. In the lower powers the lens curvatures are weaker, and as a general rule this causes a greater concentration of flare in the image plane. The 32-mm., 0.12 N.A.; 24-mm., 0.20 N.A.; and 16-mm., 0.25 N.A. all have the same number of surfaces, yet their flare values (Figs. 9, 10, and 11) are quite different due to the different lens curvatures.

It should be emphasized that all results described above were obtained with central aperture (plane glass) vertical illumination. Flares are very much lower in half-aperture prism illumination. Photoelectric readings on flare, using the cast iron specimen photographed in Figs. 8 to 18, showed less than 1 per cent flare for all objectives in the 32-mm. to 6-mm. range when using half-aperture prism illumination. The prism illuminator is of course not usable for higher powers, hence no study was made of flares for powers higher than the 6-mm. objective.

Filming the objectives increases their transmission so that a shortening of the photographic exposure is to be expected. The relative amount by which the exposure is shortened depends on the reflection characteristics of the specimen. Indeed for very dark specimens the filmed objective may require a *longer* exposure than the unfilmed, due to the flare from the unfilmed objective approaching the amount of light reflected from the specimen. In general, however, a definite and useful exposure shortening has been achieved by the filming. This should be of help to the metallurgist, particularly in the higher powers where the exposure shortening is greatest, and also most needed.

To take a single example, the 10.25-mm., 0.40 N.A. objective, with 6 air-glass surfaces, showed exposure shortenings of:

29 per cent on manganese bronze
32 per cent on cast iron
42 per cent on cast steel.

Objectives with fewer air-glass surfaces show less gain in this respect while those with more air-glass surfaces show more.

Results obtained for an objective in vertical illumination cannot of course be used in determining the effect of filming on the same objective used in transmitted illumination. In the latter case the flares are multiple reflections from the lens surfaces, and hence much weaker than the direct reflections which are the principal cause of the flares in vertical illumination. The multiple reflections occurring in transmitted illumination are of course also present as a minor source of flare in vertical illumination.

Summary

The application of anti-reflection films to metallographic objectives produces a gain in image contrast. The amount of this gain depends not only on the objective design but also on the type of specimen being observed. A definite and useful gain in contrast is obtained on specimens having a long tone range, low reflecting power, and low contrasts. Little gain in contrast is obtained on contrasty specimens of high reflecting power having few shades of gray. The basic cause of flare variations in different lens designs appears traceable principally to the relative spreading of the flares in the image plane rather than to differences in the number of air-glass surfaces in the lens system. If we accept the philosophy that all objectives

should give good crisp imagery on the "tough" as well as the "easy" specimens, filming the objectives appears advisable. The metallurgist is, however, cautioned against expecting marked improvement on contrasty specimens of high reflectivity.

Acknowledgment

The writer is indebted to Messrs. Vernon H. Patterson and Richard C. Hendrickson of the Bausch & Lomb Metallurgical Department for the preparation of metal specimens used in this study, and for advice on approaching the problem from the metallurgist's point of view.

DISCUSSION

Written Discussion: By Paul A. Beck, Department of Metallurgy, University of Notre Dame, Notre Dame, Ind.

Mr. Benford is to be congratulated on the excellent work reported in this paper. I believe that his results constitute an important advance in the construction of metallographic objectives. What makes me really enthusiastic, however, is that this is only the beginning, and that further great improvements may be expected with the development of improved anti-glare films.

Oral Discussion

T. G. Rochow:[3] First, I should like to make a prediction: In general, filmed objectives will show greater improvement in image when they are used with mineralographic and resinographic polished specimens than they will with metallographic ones. Mineralographic specimens generally have long tone ranges (e.g., pyrite, sphalerite, and quartz) and low reflecting power (sulphides and oxides, compared with elementary metals). Resinographic specimens generally have very low contrast (e.g., in brake lining there may be two resinous phases, besides cotton and chrysotile). In mixed specimens, there may be extremely long tone ranges (e.g., copper powder, graphite and melamine resin). These considerations are important because the microscopical examination of opaque minerals is already an extensive field of investigation and it is reasonable to predict that the microscopical examination of thermoreacted resins, by reflected light, will find extensive use in the near future.

I also have a question. I should like to ask Mr. Benford whether or not filming has any effect, good or bad, on the correction of chromatic or spherical aberrations in a given lens system.

D. Lewis:[4] I would like to ask how durable these magnesium fluoride films

[3]Microscopist, American Cyanamid Co., Stamford, Conn.

[4]Chief development engineer, John A. Roebling's Sons Co., Trenton, N. J.

are. Are they as hard as glass or are they softer than glass, and if they are softer, what happens when they have been wiped daily for several years?

RUSSELL G. FERNALD:[5] I would like to ask what improvement a filmed lens makes in visual examination in annealed brass. Is there a definite improvement?

J. R. VILELLA:[6] The purpose of this paper, which is to evaluate experimentally the improvement in quality of image brought about by coating microscope objectives with anti-reflection films, seems to me to have been well achieved by the series of experiments described. I have no comment to make on the optics involved in this work, and concerning the author's general conclusions I can only say that, in my opinion, if the author errs it is in being excessively modest. However, there are certain aspects of the application of anti-reflection films to microscope objectives not discussed in this paper, which are of great interest to the metallographer.

Those of us who like to make the best possible photomicrographs find that in order to disclose the finest and most delicate details of structure we have to etch the specimen very lightly. When we etch lightly the contrast between the various constituents of the structure is so slight that we are forced to make most of our photomicrographs on plates or films of very high contrast. High contrast photographic emulsions are generally of slow speed, and for this reason we have to resort to strong illuminants such as the carbon arc. This is undesirable; a carbon arc at its best is a necessary evil, for it flickers, wanders, and sometimes goes out during the exposure. To prevent large variations in intensity and to obtain even illumination, it requires close attention and frequent adjustment of the gap between electrodes. In other words, the steady source of illumination provided by incandescent filament bulbs is highly desirable in metallography, but it cannot always be used because it lacks intensity for photomicrography on slow, high contrast emulsions.

The improvement in contrast brought about by objectives and eyepieces coated with anti-reflection films seems to be sufficiently great to permit the use of high speed, lower contrast emulsions for photomicrography. This in turn would permit the use of incandescent bulbs as illuminant.

It seems to me that the coating of microscope objectives with anti-reflection films, by the very fact that it improves the image contrast between the various constituents of the structure, promises to do two distinct services to metallography, namely, to permit us to etch the structure as lightly as is desirable for revealing maximum detail and to do away with the undesirable carbon arc as a source of illumination.

Author's Reply

In reply to Mr. Beck, anti-reflection film materials and methods of making films are constantly being studied by practically every concern engaged in manufacturing optical instruments. Under such conditions it is almost inevitable that improvements will be made in the near future.

Dr. Rochow is unquestionably right in predicting that filmed objectives will show a greater improvement in the nonmetallic fields mentioned. The nonmetals

[5]Metallurgical engineer, Western Electric Co., Chicago.

[6]Research Laboratory, United States Steel Corp., Kearny, N. J.

have comparatively low reflecting powers, frequently in the neighborhood of the reflecting power of the air-glass surfaces of the objective lenses, and are invariably difficult to view in reflected illumination. Dark-field, oblique illumination, or polarized light, is frequently required to see such specimens with any degree of satisfaction.

The effect of filming on the spherical and chromatic aberrations is completely negligible. The film increases the radius of curvature and also the thickness of the lens elements by about 0.0001 mm., an insignificant amount even on the tiny lenses in high power immersion objectives.

In response to Mr. Lewis, the films are approximately the same hardness as glass. They will scratch when mistreated just as glass will. Proper treatment of lenses, filmed or unfilmed, requires that soft clean cloths be used. Harsh abrasive material should be blown or dusted off before cleaning is begun. The normal cleaning fluids (alcohol, xylol, acetone, ether, or soap and water) may be used on filmed surfaces.

Replying to Mr. Fernald, there is a definite improvement. The photographic reproduction, Fig. 5, does not show as much improvement as will be evident visually, due to inevitable loss of tone values in the photographic and half-tone processes. It should be emphasized that the visual improvement exceeds the photographic improvement.

Mr. Vilella has brought up an important point which had escaped the writer. It is certainly an advantage to go to higher speed and lower contrast film if the image has enough contrast to stand it. It is probably fair to say that a good many metallographic pictures suffer from lack of tone range due to the high contrast photographic materials normally used.

We agree with the statements on the undesirable features of the carbon arc. The tungsten ribbon filament is definitely preferable, giving steadier and more uniform illumination, and being at the same time much more convenient to operate. Our thoughts along this line had been to make the tungsten filament practical by increasing the efficiency of the vertical illuminator, and by increasing the transmission of the objectives. Mr. Vilella goes one step further and shows that the gain in contrast makes faster photographic materials usable, so that the lower intensity tungsten filament becomes a more practical source for metallography than it has been in the past. The order of magnitude of this gain in photographic efficiency far exceeds the gains we have been able to make by increasing the efficiency of the optical system.

TEMPER BRITTLENESS

By John H. Hollomon

Abstract

During the tempering of most medium and high alloy steels a transformation can occur which decreases the impact energy of the steel without material effect on the other physical properties. Previous measurements on the effects of metallurgical variables upon the susceptibility to this embrittlement have been re-evaluated. The need for new data is apparent. The available information indicates, however, that the decrease in impact energy is caused by a precipitate which forms from solid solution as do precipitates which cause age hardening. It is suggested that the precipitate may be iron nitride. A discussion of the relations between the effects of this precipitate on the impact properties and its effects on the flow and fracture characteristics of steel is included in this paper.

Introduction—Historical

RECENT studies indicate that practically all medium or high alloy steels are subject to a loss of impact energy if tempered in a certain range of temperature or if cooled slowly through this range. In recent years, little study has been given to the transformation which produces this effect. Many brittle failures of steel parts attributed to other causes are now being traced to this temper brittleness. It has been found necessary in designing or choosing the composition for such parts to understand the transformation and minimize its effects.

As early as 1883, some blacksmiths were aware that steels frequently had to be quenched from the tempering temperature to avoid embrittlement. H. M. Howe (1)[1] pointed out that this treatment of steel was called "water-annealing" by those who employed it. The

[1]The figures appearing in parentheses pertain to the references appended to this paper.

The statements or opinions in this article are those of the author and do not necessarily express the views of the Ordnance Department.

A paper presented before the Twenty-seventh Annual Convention of the Society held in Cleveland, February 4 to 8, 1946. The author, John H. Hollomon, is Captain, Ordnance Department, U. S. Army, Watertown Arsenal, Watertown, Mass. Manuscript received April 4, 1945.

early Krupp patents (2)[2] (circa 1900) indicate that armor manufacturers recognized the necessity of quenching nickel-chromium steels following tempering. About the time of World War I, Ordnance metallurgists, of the major powers at least, in studying the properties of armor and gun steels found that many nickel-chromium steels were brittle when notched and broken in bending, if furnace-cooled after the tempering treatment. They found also that if specimens from the same steel were quenched following an identical tempering treatment, the results of notched-bar tests indicated that they were ductile. This difference in behavior was not generally reflected in the results of tensile tests. Furthermore, steels susceptible to this "temper brittleness", which first was called "Krupp Krankheit" since it was encountered in Krupp armor steels, often failed brittlely in service (little energy was required for fracture). Since many cases arose in which failure of steel parts occurred and subsequent examination indicated that none of the mechanical properties, other than the energy required to break a notched-bar, were deficient, the impact test became widely used in England at least, in the acceptance testing of steels, particularly for those to be used in Ordnance materiel. Apparently as early as 1900, it was realized that such embrittled steels not only exhibited subnormal notched-bar energies, but also revealed "crystalline" fractures. On the basis of this observation, fracture tests were used in England (and possibly in France) for accepting armor. These tests consisted simply of notching specimens, breaking them across the thickness of the plates, and examining the fractures. Armor plates, which revealed nonfibrous fractures when tested in this manner, were rejected or re-treated. This information gave considerable impetus not only to the study of temper brittleness itself, but also to the study of the results of notched-bar impact tests as affected by changes in steel composition and manufacturing procedures.

The first published reference to the phenomenon of temper brittleness seems to be a discussion (3) by H. Brearley of a paper entitled "The Use and Abuse of Steel" which appeared in 1917. The authors of the paper had stated that steels and, in particular, nickel-chromium steels, could be slowly cooled after tempering. Brearley pointed out that frequently such steels must be quenched from the tempering temperature, and further that the notched-bar impact energy of these steels will deteriorate if retempered between 300 and

[2]References are only to U. S. Patents. Earlier German Krupp Patents were not available.

500 degrees Cent. (570 and 930 degrees Fahr.)[3] following quenching from an original tempering at 650 degrees Cent. (1200 degrees Fahr.). He further related this same phenomenon to the abnormally low values of impact energy often obtained with nickel-chromium steels when tempered in the range of 300 to 500 degrees Cent. (570 to 930 degrees Fahr.) directly after the quench. Brearley implied that the embrittling arising either from slow cooling through the range 300 to 500 degrees Cent. (570 to 930 degrees Fahr.), reheating into this range following a higher temperature tempering treatment, or retempering in this range following quenching after austenitizing were all related and occurred more frequently with (acid) open-hearth than with crucible steel. Grenet (4), in France, who appears to have studied the phenomenon independently, arrived at very nearly the same conclusions. An extremely interesting discussion of the manufacture of ferrous ordnance materials in England during the first World War (including references to temper brittleness) is to be found in Brearley's autobiography: "The Knotted String" (5). Other early references to the phenomenon of temper brittleness occur in the papers and discussions of Hatfield (6), Philpot (7), and Dickenson (8). These authors appear to have based their comments primarily on the experiences in the production of armor plate in England and on the results of an experimental program carried out at Woolwich Arsenal.

Manufacturers of ordnance steel in America were also aware of temper brittleness although there appears to have been little or no exchange of information between workers in England and America. Langenberg (9) of Watertown Arsenal presented the results of tests on several nickel-chromium steels which were temper brittle. Early experiments performed during World War I by Wood and Archer (10) were presented by Jeffries (11) in his paper entitled "Physical Changes in Iron and Steel", which did not appear until 1922.

The first paper (12) discussing the comprehensive research performed at Woolwich during the first World War was written by R. H. Greaves and published in 1919. The paper is the first of several (13), (14), (15), (16) by Greaves and his collaborators and are still the most definitive that have been written. The ideas expressed by Brearley were discussed and expanded, the effects of alloying elements and other metallurgical variables were studied and a standard test for measuring the relative susceptibility to the embrit-

[3]Temperatures in degrees Centigrade are used primarily throughout the paper since this temperature scale is used in practically all papers which discuss temper brittleness.

tlement developed. Since, as will be discussed, the method of measuring this relative influence is so important to the interpretation of the results, the exact procedure which has been used is described below.

"Two similar samples of the steel to be examined are oil-hardened from 900 degrees Cent. (1650 degrees Fahr.), and tempered for 2 hours at 650 degrees Cent. (1200 degrees Fahr.). One sample is cooled in water after tempering, the other is cooled slowly, the rate of cooling between 600 and 400 degrees being uniform and 0.3 degree Cent. per minute. In cases of slight susceptibility further investigation is made by raising the hardening temperature to secure better discrimination after quick and slow cooling from the tempering temperature" (14). Specimens of each treatment were notched and broken by an impact blow.

It is to be noted that in describing the test nothing is said concerning the structure the steel should possess on quenching, or whether in forgings the specimens should be taken parallel or perpendicular to the forging direction, even though Greaves and Jones recognized the importance of such factors (14). Furthermore, almost all notched-bar impact tests for temper embrittlement susceptibility have been performed at room temperature. The relative effects of the various alloying elements on the susceptibility of steels to temper brittleness were measured by determining their effects on the ratio of the impact energy of the water-quenched to the furnace-cooled specimens. As measured in this manner, nickel, chromium, manganese, and phosphorus were all found to increase the susceptibility toward temper embrittlement, while molybdenum was found to decrease it. Greaves and his collaborators pointed out the effect of increasing the austenitizing temperature in increasing the susceptibility ratio and noted that steel making practice also appeared to influence the results. These investigators showed that the transformation which was responsible for the embrittlement occurred in about the range of from 400 to 600 degrees Cent. (750 to 1110 degrees Fahr.) (for reasonable times) and could occur on tempering in this range, on cooling from temperatures higher than about 600 degrees Cent. (1110 degrees Fahr.) or on reheating in this range, following quenching from tempering temperatures higher than about 600 degrees Cent. (1110 degrees Fahr.). They concluded that the transformation which occurred in this region was a simple precipitation and suggested that the precipitate might be chromium oxide.

Many metallurgists during this early period attempted to find some manifestation of the transformation other than the decrease in the energy required to break a notched bar. A slight change in density and a slight evolution of heat (17) were both believed to have been observed with susceptible steels following or during treatment in the critical temperature range. Many early investigators studied the microstructure of brittle and ductile specimens and Andrew and Dickie (17) and Haskell and Mann (18) presented a pair of photomicrographs purporting to show the presence of a precipitate at the grain boundaries in embrittled specimens.

Since World War I a number of metallurgists have published the results of researches concerned with temper brittleness (see following sections and bibliography), but they contain little fundamental information in addition to that described in the English papers published in the early 1920's. The importance of the embrittlement of steels used in oil refineries and turbines for operation in the range of temperature of about 400 to 550 degrees Cent. (750 to 1020 degrees Fahr.) has been discussed by many workers in both England (19), (20) and America (21). Investigators of this problem studied the effects of very long time treatments in the embrittlement range on steels being considered for these applications.

An apparently new effect was noted by Houdremont and Schrader (22) and by Bischof (23), who found that retempering a number of times (above about 650 degrees Cent.) reduced the susceptibility ratio. Jolivet and Chouteau (24) found similarly that increasing the time or temperature of tempering (above 650 degrees Cent. but below the Ac_1 temperature) also reduced the susceptibility ratio. Also certain rather recent work by Lea and Arnold (25) indicates that molybdenum possibly only retards the transformation which causes temper brittleness and does not prevent it. These investigators also found that tensile specimens of severely embrittled steels fractured longitudinally rather than with the normal transverse cup-cone fractures. This type of fracture has been observed by others (26).

Recent studies at the Watertown Arsenal Laboratory again have emphasized the importance of the problem to development of steels having the maximum impact properties consistent with their hardnesses. New information and a new interpretation of the mechanical behavior of metals indicated that not only was a re-evaluation of the knowledge concerning temper brittleness desirable, but it was also necessary.

The effects of the transformation on the properties of susceptible steels will first be discussed in detail since the interpretation of the effects of variables which are discussed in the subsequent sections depend to a large extent on an understanding of the methods used to measure their effects.

Effect of the Transformation

A. *On Impact and Tensile Properties*—In the early studies of temper brittleness it was found that the primary effect of holding

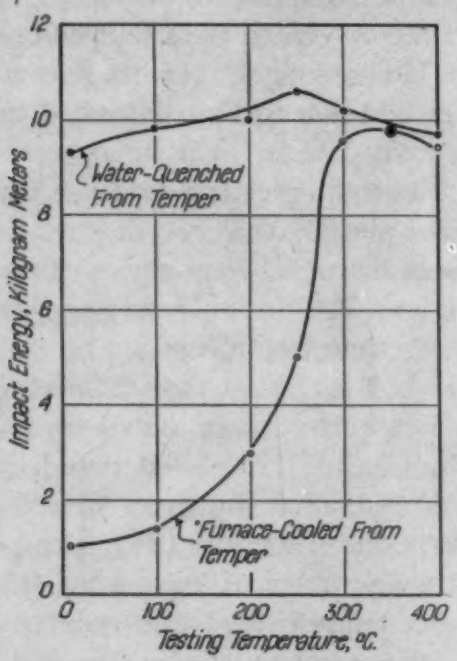


Fig. 1—Impact Tests at High Temperatures of a Susceptible Nickel-Chromium Steel (After Monypenny).

susceptible steels in the temperature range of about 450 to 600 degrees Cent. (840 to 1110 degrees Fahr.) was to lower the energy required to break notched-bar specimens in impact at room temperature. As will be seen in the following sections, no tests other than the notched-bar test (with the possible exception of microscopic studies) have been able to differentiate effectively between embrittled and nonembrittled specimens.

Even though most notched-bar tests for temper brittleness susceptibility have been performed at room temperature, Monypenny (27) in 1920 presented impact curves for a range of testing tempera-

tures of a susceptible steel in the embrittled and ductile conditions. His curves are reproduced in Fig. 1. Based upon Monypenny's remarks it can be inferred that the specimens consisted of the martensitic structure before tempering. These curves, as well as several to be presented subsequently, reveal a very disturbing situation. The ratio of the impact energies of water-quenched and furnace-cooled specimens tested at 300 degrees Cent. (570 degrees Fahr.) is about 1, at 200 degrees Cent. (390 degrees Fahr.) it is about 3 and at 20 degrees Cent. the ratio is approximately 9. Thus, the susceptibility ratio varies with the severity of the impact test[4].

Since the temper brittleness transformation causes a shift in

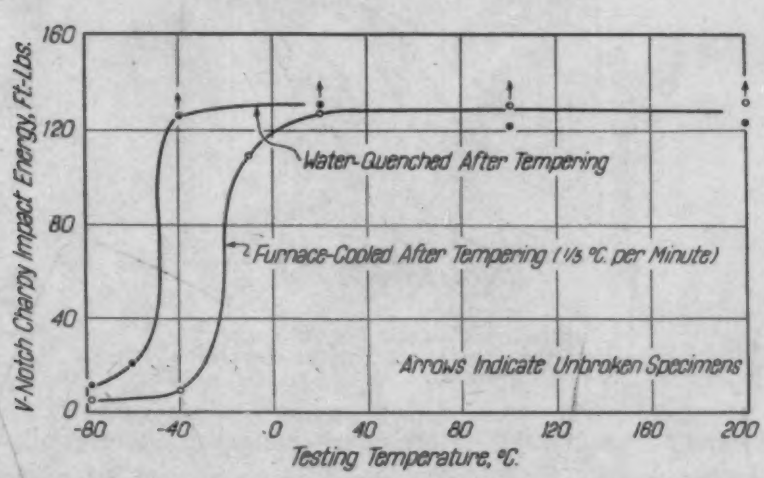


Fig. 2—The Effect of Temper Brittleness on Impact Properties of a Manganese-Vanadium Steel (Bainite Structure on Quenching—Tempered at 650 Degrees Cent.). (Appendix A.)

the impact curves, the range of testing temperatures over which the temper brittleness is evident will depend upon the properties of the nonembrittled specimens.[5] For example, if the structure of the steel had been tempered bainite rather than martensite (and the impact test had been performed at lower temperatures) the impact curve of the nonembrittled specimens would have had a form similar to that illustrated in Fig. 2. The impact energies of specimens of a manganese-vanadium steel (see Appendix A) having a (tempered) bainitic structure are presented in this figure in both the water-quenched and furnace-cooled conditions (after tempering). Had the structure of the steel been other than bainite on quenching the impact properties

[4]A similar situation would have existed had specimens of various notch conditions been used and the impact energy plotted as a function of notch severity.

[5]In a recent paper Mikhailov-Mikheev (28) alludes to this problem.

of the nonembrittled specimens would have been different and the susceptibility determined by comparison of impact energies obtained at one testing temperature would have been changed. Thus the *apparent* susceptibility as a result of tests at 20 degrees Cent. may vary even though the embrittling treatments are identical (and the steel is the same) if the "as-quenched" structure is changed.

The impact curves of nonsusceptible or nonembrittled susceptible steels depend upon the structure of the steel and its "quality".[6] As illustrated schematically in Fig. 3, the impact curves of tempered steels of the same "quality" shift in the direction A to E as the

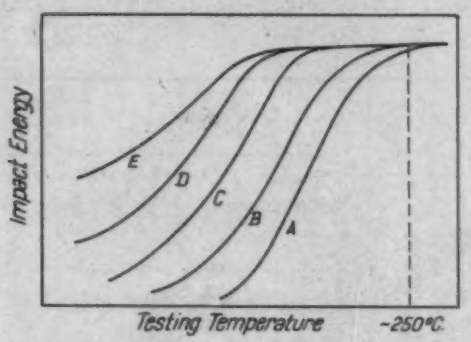


Fig. 3—Schematic Impact Curves for Tempered Steels of Various Structures (Equal Hardness).

quenched structures change from coarse pearlite through bainite to tempered martensite. The impact curves as a function of temperature also depend upon the degree of tempering. The transition shifts toward lower temperatures and the level of the impact energy rises as the hardness is decreased by tempering. Impact curves of a pearlitic steel after different tempering treatments are illustrated schematically in Fig. 4. The curves shift from A to D as the hardness (yield strength) decreases by tempering. The effect of temper brittleness can be imposed on any of these impact curves as illustrated in Figs. 1 and 2; the degree of the embrittlement depending on the amount of transformation which occurs.

Thus, for a given steel, the susceptibility ratio (the apparent temper brittleness) depends upon the testing temperature and the impact properties of the nonembrittled steel as well as upon the amount of embrittlement. The impact properties of the nonembrittled specimens depend in turn upon the "quality" of the steel and on its structure.

[6]The effects of manufacturing variables: casting, refining, etc.

Relatively few tensile tests have been performed to compare the properties of embrittled and normal specimens. Greaves and Jones (15) performed some tests on specimens water-quenched and furnace-cooled (⅓ degree Cent. per minute) from tempering temperatures above 600 degrees Cent. (1110 degrees Fahr.) and found no difference in the normal tensile properties. Greaves and Jones observed, however, that the stress-strain curves deviated from linearity sooner for the quenched specimens than for the slowly cooled and attributed this difference to residual stresses introduced by the rapid cooling. Philpot (7) gives several examples of comparisons

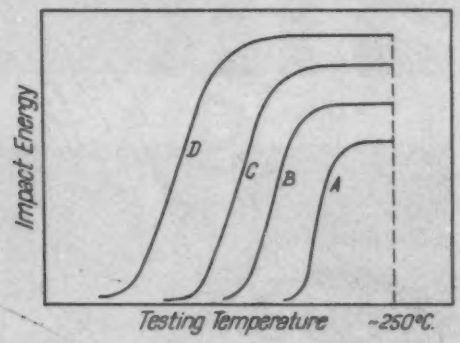


Fig. 4—Schematic Impact Curves for a Pearlitic Steel Tempered to Several Hardnesses.

between the tensile properties of water-quenched and furnace-cooled specimens of very susceptible steels in which the reductions of area were considerably less in the embrittled than in the normal conditions.[7] However, the rate of furnace cooling used for all these tests was not very slow and the embrittlement developed was not as severe as can be produced by holding for long times at temperatures below about 500 degrees Cent. (930 degrees Fahr.).

Lea and Arnold (25) observed a very interesting effect of severe embrittling treatments on the fracture of tensile specimens of nickel-chromium steels. Several other investigators have observed this peculiar type of fracture at low testing temperatures with steels tempered in the embrittlement range. Lea and Arnold, however, performed their tests at 20 degrees Cent. Pictures of tensile fractures of severely embrittled steels are illustrated in Fig. 5. The specimens treated for long times at 500 degrees Cent. (930 degrees Fahr.) fractured longitudinally under the action of load applied parallel to the axis of the specimen. Pictures illustrating the varia-

[7]Some of his results show no differences, however.

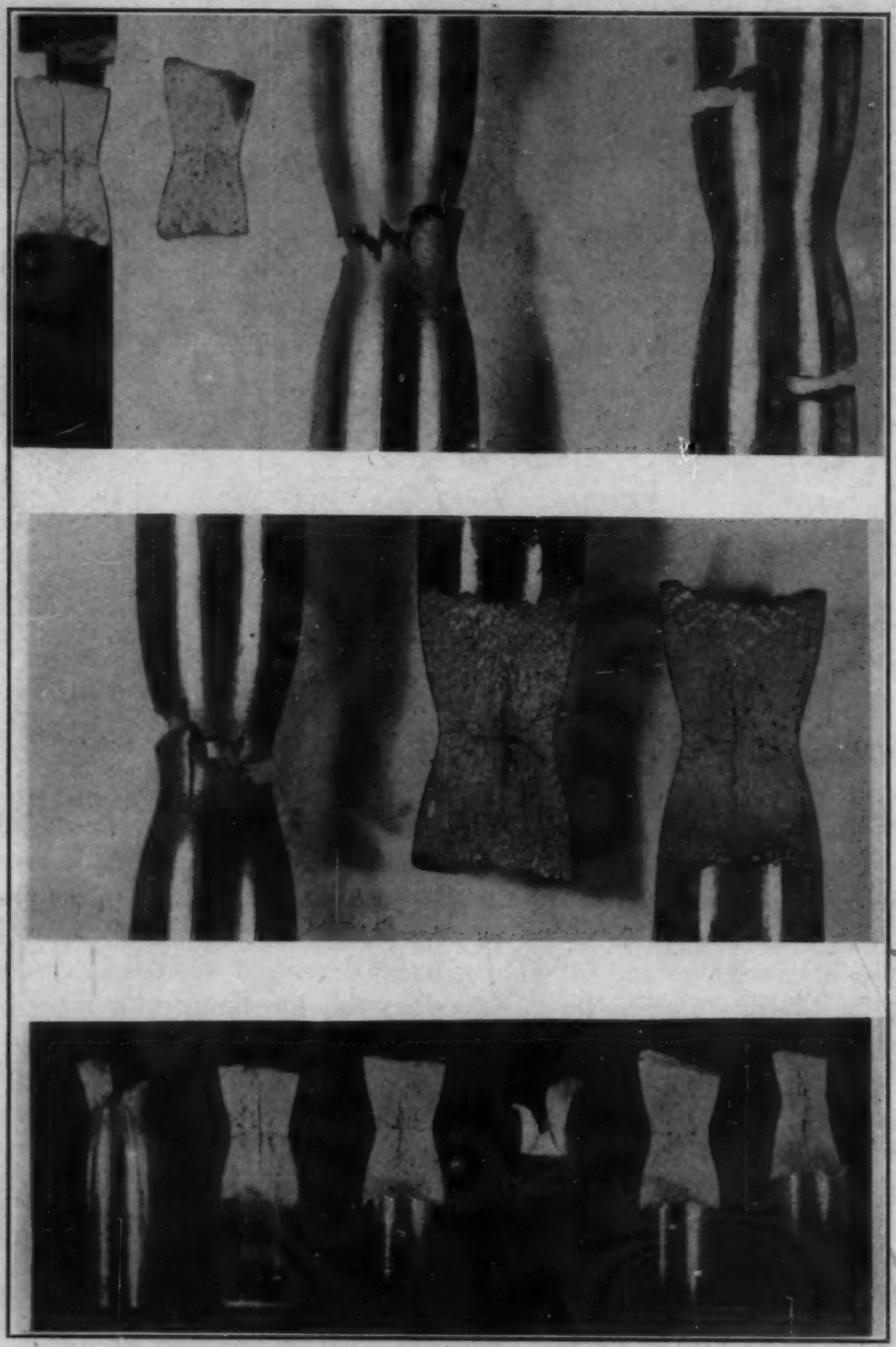
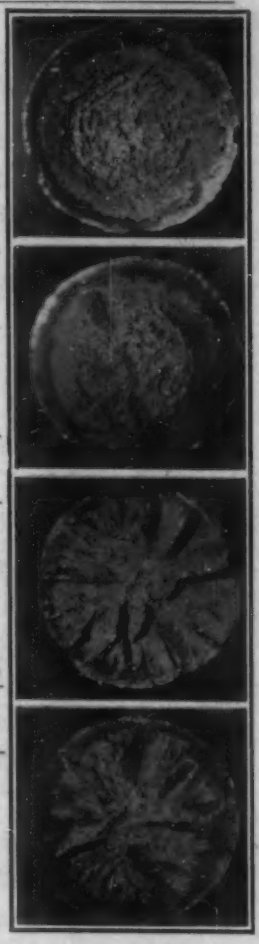


Fig. 5—Longitudinal Fractures of Embrittled Specimens (from Lea and Arnold).

Time at 480 Degrees Cent.	Yield Strength 0.01% Offset (1000 Psi.)	Yield Strength 0.1% Offset (1000 Psi.)	Tensile Strength (1000 Psi.)	Reduction of Area (Per Cent)	Fracture
0	127.5	134.4	146.5	63.7	
½ hour	130.0	133.8	145.0	64.4	
5 hours	121.3	133.8	145.0	62.7	
50 hours	130.0	136.9	151.0	60.6	

All specimens treated in ⅝-inch squares austenitized at 925 degrees Cent., tempered at 660 degrees Cent. and embrittled at 480 degrees Cent. for indicated times.

Fig. 6—Result from Tensile Tests of Susceptible Manganese-Molybdenum Steel.

tion of the type of fracture with increasing embrittlement of a manganese-molybdenum steel are shown in Fig. 6 (see Appendix A for composition). In this case the fracture changed only from the normal cup-cone to a star-type fracture.

B. *On Other Physical Properties*—Many properties other than tensile and impact have been investigated in an attempt to understand the mechanism of temper brittleness. Greaves and Jones (14) pointed out in an early paper that strains were introduced on quenching from high tempering temperatures and that the effect of stress relief must be considered in interpreting the measurements of physi-

cal properties following heating after quenching. Greaves and Jones also pointed out that specimens in the tough and brittle conditions must receive exactly the same amount of tempering if their properties are to be compared. Such comparisons have been made by first slowly cooling one specimen and water quenching another (from above 600 degrees Cent.), reheating both specimens and slowly cooling the one that had been quenched and quenching the one that had been slowly cooled. When this procedure is followed the embrittled and ductile specimens of susceptible steels should differ only in the transformation which is the cause of temper brittleness and in the magnitude of residual stresses.

On using the above described technique, Greaves and Jones (14)[8] found no consistent difference in hardness between tough and brittle specimens. They did find, however, a difference in specific gravity of from 0.0004 to 0.0010 between such specimens but also found that this difference was greatly reduced if the tough steel were reheated to 200 degrees Cent. (390 degrees Fahr.). This difference in specific gravity Greaves and Jones attributed to residual stresses, and its reduction by heating to 200 degrees Cent. (390 degrees Fahr.) was believed to be due to the partial relief of the stresses. They concluded that the difference in density due solely to the transformation which is the cause of temper brittleness appeared to be less than one part in 20,000.[9]

Andrew and Dickie (17) found changes of hardness and specific gravity of steels as a function of cooling rate after tempering. These experiments are subject to the criticism that the specimens were probably tempered at temperatures above the critical. (See page 485.)

No appreciable differences[9] between the tough and brittle conditions have been reported to have been found by electrical resistance, magnetic, dilatometric or thermo-electric measures even though these techniques have been used (29) in temper brittleness studies.

Rogers (30) showed differential heating curves which indicate that possibly an evolution of heat occurs on heating after quenching (from the austenitizing temperature). A great deal of discussion was incited by these results of Rogers, and Greaves and Jones (31) state that "the application of his results to the question of temper brittleness is by no means clear", because of the effects of other vari-

[8] A summary of studies on physical properties other than tensile and impact may be found in a paper by Andrew and Dickie (17).

[9] It must be pointed out, however, that none of these physical tests appear to have been performed on specimens severely embrittled by holding at low temperatures.

ables which Rogers did not take into account. Greaves and Jones (14) themselves state that they could find no *consistent* evidence of a transformation by means of thermal techniques.

C. *On Metallurgical Structure*—Many investigators have attemped to discover the alteration in microstructure which is responsible for the embrittlement. Grenet (4), for example, stated in 1919 that he had not been able to differentiate tough and brittle specimens by microscopic methods. Apparently the first published reference to microscopic work in which a difference was observed is to be found in the paper by Andrew and Dickie (17). They state that they had observed a grain boundary precipitate in several (but not all) embrittled steels and show a single photomicrograph illustrating the precipitate. These investigators, however, heated their specimens above the eutectoid temperature before slowly cooling. Carbides would be expected to precipitate in any steel following such treatment. The relation of these structural studies to the problem of temper brittleness is therefore not clear.

A year after the paper by Andrew and Dickie was published Haskell and Mann (18) in a paper on high velocity impact testing presented a photomicrograph which they state shows a grain boundary precipitate in temper embrittled specimens (Murakami etch). Other unpublished work at Watertown Arsenal during this period revealed this precipitate in several embrittled steels. Haskell and Mann inferred that the precipitate was a carbide. Recently at the Watertown Arsenal Laboratory, Cohen, Hurlich, and Jacobson (32) have obtained photomicrographs of water-quenched and furnace-cooled specimens of susceptible steels after etching in a special reagent which indicate that a precipitate forms at the prior austenitic grain boundaries during furnace cooling. It has not yet been shown definitely that this precipitate is connected with temper brittleness but further work is in progress. These investigators have also found that reagents which delineate carbides do not reveal the grain boundary precipitate, and they deduce from their etching experiments that the precipitate is at least not an ordinary carbide.

The Effect of Variables

A. *Time and Temperature of Transformation*—1. *Isothermally*—It must be concluded from the discussion of preceding sections, that the only available tool for measuring quantitatively the ex-

tent of the transformation which results in temper brittleness is the notched-bar impact test. Even though the results of such measurements are very deceptive when different steels or steels of different structures are tested, some information regarding the nature of temper brittleness may be obtained by careful examination of the available data.

As indicated in the introduction, the transformation which results in temper brittleness occurs only below about 600 degrees Cent. (1110 degrees Fahr.). The transformation is reversible. For example, if a steel has been embrittled by slowly cooling from above 600 degrees Cent. (1110 degrees Fahr.) its properties may be restored by heating to above 600 degrees Cent. (1110 degrees Fahr.)

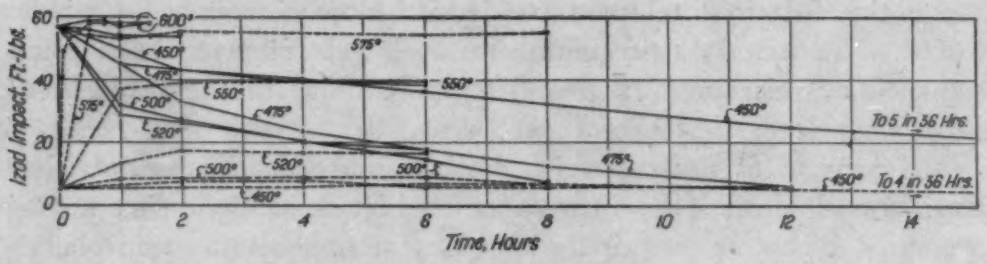


Fig. 7—The Effect of Reheating Brittle and Ductile Specimens of a Susceptible Steel Between 450 and 600 Degrees Cent. (After Greaves and Jones; Their Steel D).

and cooling rapidly. Greaves and Jones (14) prepared a large number of specimens of a susceptible steel, heated them to 650 degrees Cent. (1200 degrees Fahr.) (after quenching to martensite), quenched them in water to prevent embrittlement and then reheated them to various temperatures between 450 and 600 degrees Cent. (840 and 1110 degrees Fahr.) for various times. Typical curves obtained by these workers are reproduced as Fig. 7.

It can be assumed for purposes of study that when the decrease in impact value of two specimens of the same steel held for different times at two temperatures below 600 degrees Cent. (1110 degrees Fahr.) is the same, then the amount of transformation which has occurred is the same. In Fig. 8 curves are shown relating the time and temperature required to produce a given drop in impact energy. A striking similarity is apparent between these curves and those utilized to illustrate the progress of the decomposition of austenite. At temperatures just below about 600 degrees Cent. (1110 degrees Fahr.) and at very low temperatures, transformation occurs only very slowly, while at intermediate temperatures the rate is more

rapid. The significance of this information is discussed in detail in the last section of this paper.

At temperatures just below about 600 degrees Cent. (1110 degrees Fahr.), the impact energy after a very small decrease soon reaches a constant value indicating that at least for this steel the amount of transformation which occurs at temperatures slightly below 600 degrees Cent. (1110 degrees Fahr.) is small. As the temperature decreases the total decrease of impact energy (and the total transformation) increases continuously. Impact energy for nickel-chromium steels as a function of time at low temperatures are also

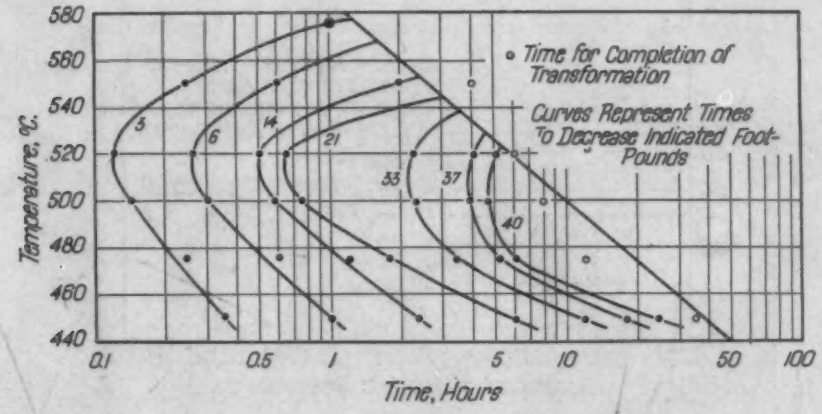


Fig. 8—Time-Temperature-Transformation Diagram for Temper Brittleness Transformation (From Data of Greaves and Jones; Their Steel D).

presented in papers by Bailey and Roberts (19) and Lea and Arnold (25).

Greaves and Jones utilized a very excellent technique for finding the time required to reach the approximate equilibrium at each temperature of supercooling. They reheated severely embrittled as well as ductile specimens of the same steel in the range of 450 to 600 degrees Cent. (840 to 1110 degrees Fahr.) and found that the impact energy of the originally ductile specimens was impaired and that of the originally brittle specimens improved. At each temperature the impact energies approached each other as indicated in Fig. 7. The value of impact energy at which the two curves coincide (within the accuracy of measurement of $\pm$ 1 to 2 foot-pounds) is a measure of the total amount of transformation which can occur at each temperature. The transformation is reversed in the severely embrittled specimens and is allowed to proceed in the originally ductile specimens. The time required for the originally ductile speci-

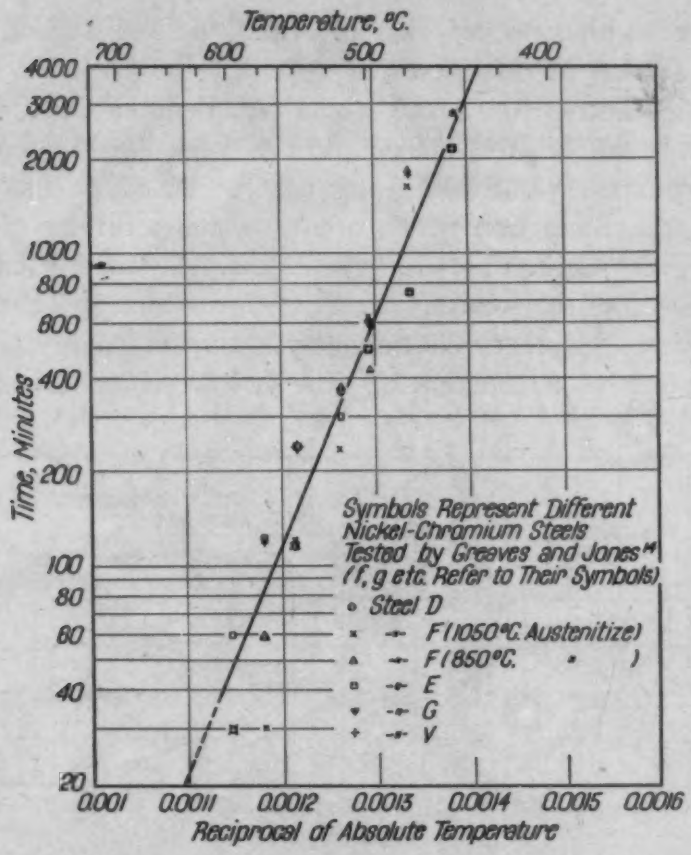


Fig. 9—Relation Between Time and Temperature of Treatment Necessary to Produce Equilibrium Temper Brittleness Transformation.

mens to reach a constant value (within ± 1 foot-pound) is taken as the completion or equilibrium time.

At the higher temperatures even though the embrittling reaction proceeds only very slowly, the time required for completion is relatively short. In fact the time required for completion increases continuously with decreasing temperatures as illustrated in Fig. 9 for several nickel-chromium steels studied by Greaves and Jones.[10]

As pointed out above, as the amount of transformation increases the impact properties decrease continuously; this is accomplished by a shifting of the impact curves as illustrated in Fig. 10. Fig. 11 illustrates the effect of increased transformation on specimens from a rolled plate of manganese-molybdenum steel. These curves were obtained by embrittling for various times at 450 degrees Cent. (840 degrees Fahr.) (Appendix A).

[10]Over the narrow temperature range it is immaterial whether log t is plotted as a function of the absolute temperature or the reciprocal of the absolute temperature. Both ways of plotting give straight lines.

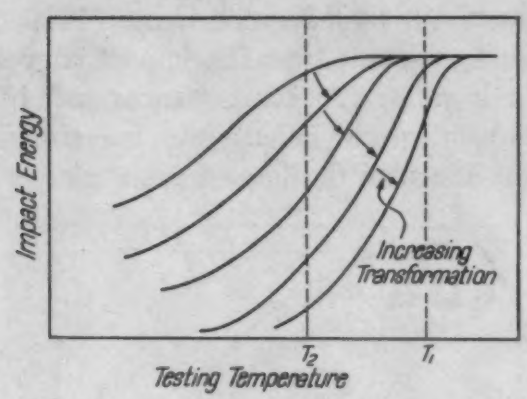


Fig. 10—Effect of Increased Embrittlement on Impact Curves.

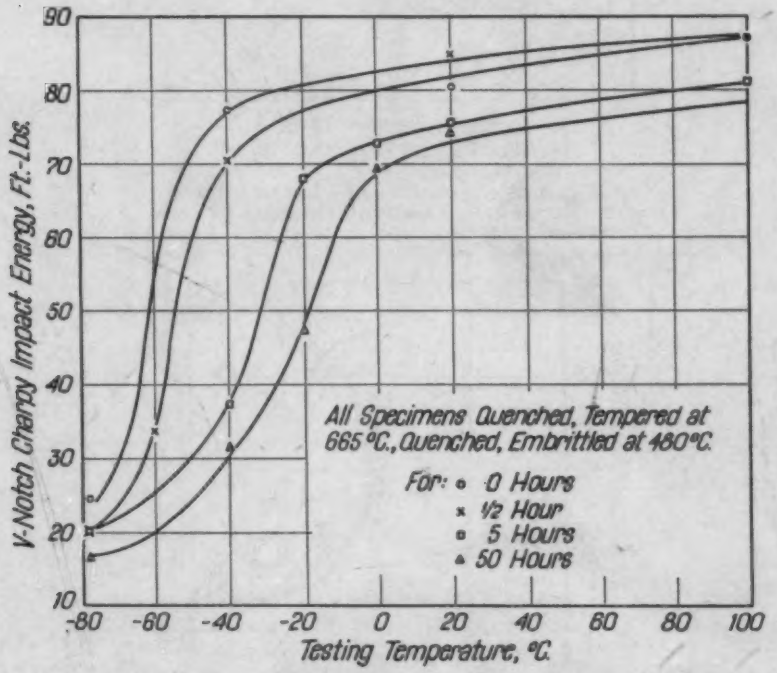


Fig. 11—Effect of Embrittlement on a Manganese-Molybdenum Forged Steel (Tempered Martensite). (See Appendix A.)

If, now, the testing temperatures with reference to the schematic curves of Fig. 10 had been T_1 and T_2, the resulting curves of impact energy as a function of increased transformation would have the form illustrated in Fig. 12. The similarity of the schematic curve illustrated for the testing temperature T_2 to the curves obtained by Greaves and Jones (Fig. 13) for the variation of the "equilibrium" impact energy with decreasing embrittling temperature is apparent.

If, however, equilibrium is not reached but the treatments at

temperatures below about 600 degrees Cent. (1110 degrees Fahr.) are carried out for a constant time the impact curves will have the form illustrated in Fig. 14. At temperatures just below about 600 degrees Cent. (1110 degrees Fahr.) the transformation proceeds slowly and the total effect on the impact properties is small. As the

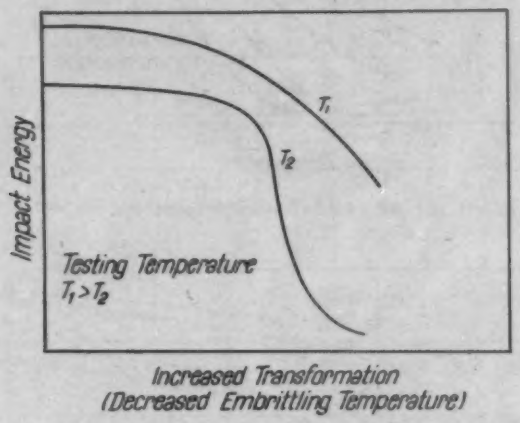


Fig. 12—Variation of Impact Energy with Extent of Temper Brittleness Transformation.

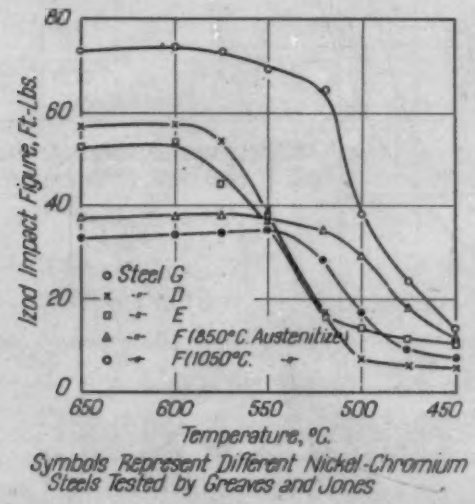


Fig. 13—Equilibrium Impact Figure (Room Temperature Tests) as a Function of Embrittling Temperature. (After Greaves and Jones.)

temperature is lowered the amount of equilibrium transformation increases but at very low temperatures the transformation cannot occur to a great extent unless the time of holding is very long. The impact properties at the lower temperatures of testing first decrease then increase as the temperature of transformation is lowered and the time kept constant. This effect of undercooling may best be il-

lustrated by a series of curves obtained by replotting some data of Greaves and Jones (14). In Fig. 15 the impact energy of a nickel-chromium steel is plotted as a function of temperature of reheating (after quenching from a tempering temperature of 650 degrees Cent.) (1200 degrees Fahr.). For short times at low temperatures

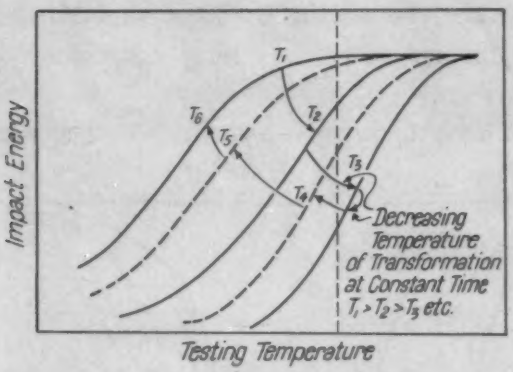


Fig. 14—Effect of Transformation at Several Temperatures for Constant Time.

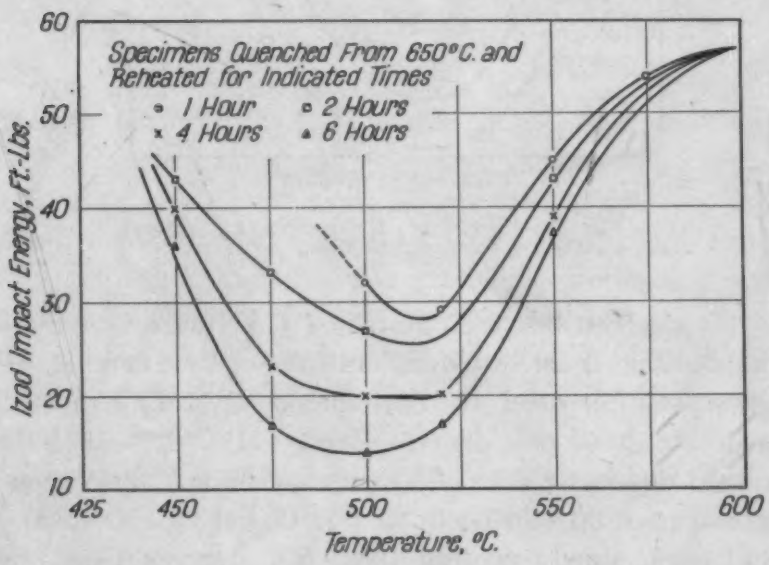


Fig. 15—Effect of Reheating Susceptible Steel in the Embrittling Range. (After Greaves and Jones; Their Steel D.)

there is no effect of the transformation because the rates of transformation are too small. As the time is increased the temperature range at which the maximum effect occurs shifts toward lower and lower temperatures until for very long times the curves approach those illustrated in Fig. 13 for equilibrium. If the impacting conditions had been more severe (temperature lower, etc.), then the curves

would have been shifted as illustrated schematically in Fig. 16. Curves similar to those of Fig. 15 have been obtained by Jolivet and Chouteau (24) and Nagasawa (33).

Houdremont and Schrader (22) and Maurer, Wilms and Kiessler (38) obtained curves showing the effect of long time treatments in the embrittling range. After the first rather rapid fall the impact energy rises very slowly with time. This very slow improvement in impact properties occurs for times long compared to the times required for equilibrium.

2. *On Continuous Cooling*—Specimens of a steel susceptible to

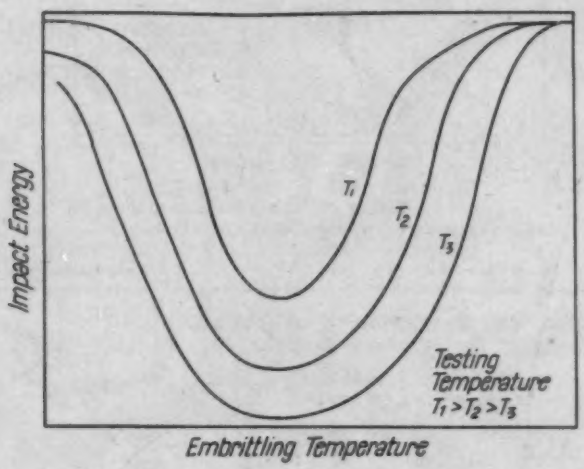


Fig. 16—Effect of Embrittling at Various Temperatures for a Constant Time.

temper brittleness can become embrittled (or further embrittled) on continuous cooling from any temperature. For example, if specimens of a susceptible steel are first quenched from tempering temperatures in excess of 600 degrees Cent. (1110 degrees Fahr.), reheated to 500 degrees Cent. (930 degrees Fahr.), and then slowly cooled, transformation will occur at 500 degrees Cent. (930 degrees Fahr.) and upon slowly cooling from 500 degrees Cent. (930 degrees Fahr.). If, however, specimens are slowly cooled from above about 600 degrees Cent. (1110 degrees Fahr.) then the embrittling will arise only from transformation on cooling. The effect of slow cooling is to cause transformation at all temperatures below the critical (600 degrees Cent.) in an amount determined by the rate of transformation and the rate of cooling. Since the impact energy is only an indirect measure of the amount of transformation which occurs, the principles for transforming the isothermal data to con-

tinuous cooling have not been formulated. If the cooling rate is sufficiently rapid the transformation is prevented in the same way the pearlite formation is prevented. As the cooling rate is decreased the amount of transformation (as measured by impact tests) increases and the impact curves are shifted in a manner similar to that shown in Fig. 10 for decreased temperatures of transformation. The same steel having different impact properties after quenching from the temper may reflect the change in cooling rate in a different fashion. The difference is reflected in the data of Jolivet and Chouteau (24) (Fig. 17). Specimens taken from a single steel were quenched

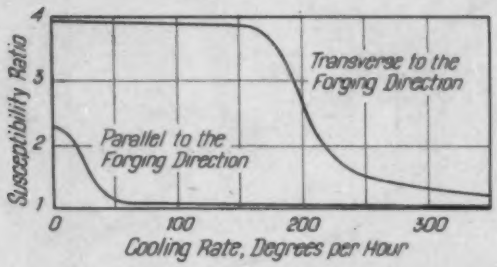


Fig. 17—Effect of Cooling Rate on the Impact Properties of a Temper Brittle Steel. (After Jolivet and Chouteau.)

to martensite (presumably), tempered at 675 degrees Cent. (1250 degrees Fahr.) for 1 hour and cooled at various rates. The impact specimens were broken at room temperature. The cooling rate at which the transformation begins to affect the impact properties is about 250 degrees Cent. (480 degrees Fahr.) per hour for the specimens taken transverse to the forging direction and about 25 degrees Cent. per hour for the longitudinal specimens. The cause of this difference in apparent susceptibility is a result of the change in the impact properties of the nonembrittled specimens. If the impact tests had been performed at lower temperatures the critical velocity would have been higher for both the transverse and longitudinal specimens.

3. *Combined Effect of Tempering and Embrittling Transformation*—As discussed originally by Brearley the transformation which is the cause of the lowered impact energy can occur during heating below about 600 degrees Cent. (1110 degrees Fahr.) following quenching (from the austenite region) as well as following tempering above 600 degrees Cent. (1110 degrees Fahr.). In the former case the agglomeration of carbides and the transformation which is the cause of temper brittleness occur simultaneously. The effect of

the transformation will depend upon the metallurgical structure of the steel and its history as well as upon the extent of the transformation. The extent of transformation in turn depends upon the time and temperature at which the tempering occurs. This concomitant effect of tempering and temper brittleness probably can be illustrated best by the data of Greaves and Jones. In Fig. 18 the impact properties of a nickel-chromium steel tempered for various times at temperatures up to 600 degrees Cent. (1110 degrees Fahr.) are pre-

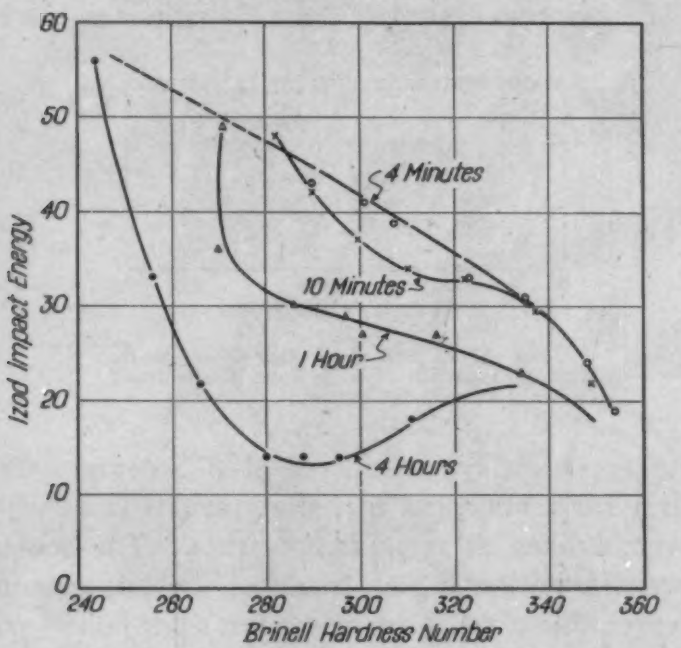


Fig. 18—Effect of Tempering at Various Temperatures for Indicated Times Followed by Water Quenching (Susceptible Nickel-Chromium Steel Presumably Martensitic When Quenched). (From Greaves and Jones.)

sented. The specimens tempered for only 4 minutes exhibit an almost continuous increase of impact energy with decreasing hardness, for the time is so short that no transformation other than normal tempering occurs. The impact energy of the specimens tempered for 1 hour is increased because the hardness decreases, but is decreased because of temper brittleness occurring at tempratures below 600 degrees Cent. (1110 degrees Fahr.). The minimum in the impact curve of the specimens tempered for 4 hours corresponds to tempering at the "nose" of the "C" curve. The specimen tempered to 245 BHN by a 4-hour treatment was heated to a temperature above 600 degrees Cent. (1110 degrees Fahr.) and hence is not embrittled.

This point lies on the smooth curve drawn through the 4-minute points. It must be remembered that the specimens having the same hardnesses were not tempered at the same temperatures for the tempering times varied. This combined effect of softening and embrittlement can be understood by a study of the curves of Fig. 19. For long tempering times the effect of the transformation will be manifest at low tempering temperatures because at low temperatures the amount of transformation possible is large. As the time of temper-

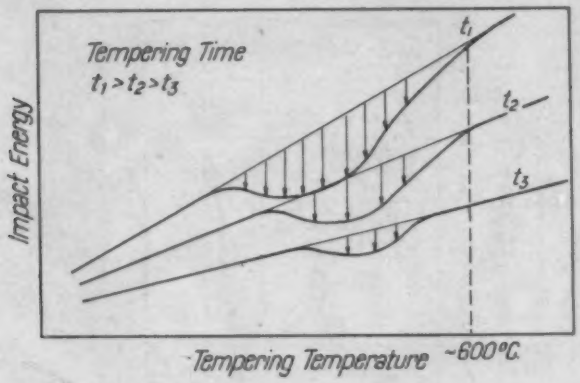


Fig. 19—Combined Effects of Tempering and Temper Embrittlement on Impact Energy.

ing is decreased the nonembrittled impact properties will decrease for each tempering temperature (because of the higher hardness) and the embrittling effect at the low temperatures will disappear. For very short tempering times the transformation can only occur at the nose of the "C" curve (if at all) and the amount which can occur is very small.

The curves given in Fig. 19 were constructed to illustrate the behavior of a steel when quenched following tempering. If, however, the specimens were slowly cooled following the temper the specimens tempered above about 600 degrees Cent. (1110 degrees Fahr.) would also become embrittled due to the slow cooling through the embrittling range. Typical curves of this type obtained by water quenching and furnace cooling after tempering are shown in Fig. 20.

B. *Time and Temperature Above 600 Degrees Cent. (1110 Degrees Fahr.)*—Houdremont and Schrader (22), Bischof (23), and Mikhailov-Mikheev (28), found that if a steel was reheated a sufficient number of times to 650 degrees Cent. (1200 degrees Fahr.) and furnace-cooled the susceptibility to temper brittleness as measured by room temperature impact tests would disappear. Houdre-

mont and Schrader reported only a slight decrease in hardness arising from this increased tempering but this statement has been questioned (34). The effect of multiple reheatings can be very simply explained by the improvement in impact properties brought about by the increased tempering (softening). This explanation is substantiated by the results of Jolivet and Chouteau (24) which are discussed below. The schematic curves of Fig. 21 illustrate the mechanism involved. As the level of impact energy is increased by

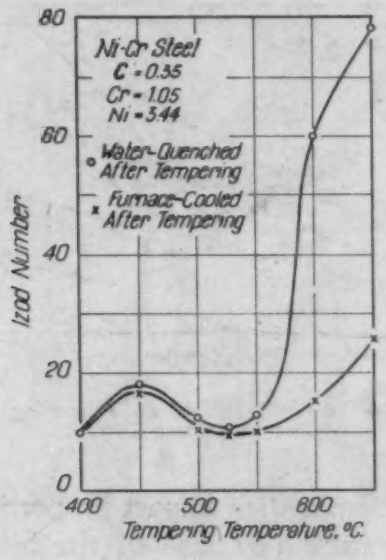


Fig. 20—Impact Properties of a Temper Brittle Steel. (After Nagasawa.)

tempering, the effect of a given amount of transformation is less evident in room temperature impact tests; the *apparent* temper brittleness susceptibility is decreased. This explanation does not say that there is *no* effect of repeated tempering but only that there *may* not be any real effect.

The data of Jolivet and Chouteau (24) if properly interpreted, however, indicate that there is no effect of increased temperature (above 600 degrees Cent.) or time at temperature other than the indirect effect attributable to softening. These investigators tempered specimens at various temperatures from 650 to 720 degrees Cent. (1200 to 1330 degrees Fahr.) for various times, quenched them and retempered them at 500 degrees Cent. (930 degrees Fahr.) for 24 hours for embrittling. The specimens were then notched and broken in impact. Their data are presented in Fig. 22 and indicate that the

impact energy increases with increasing temperature (above 650 degrees Cent.) and time. Jolivet and Chouteau did not include the hardnesses of their specimens among their results. However, it has been found (35) that the relation between time and temperature of tempering which will produce a given hardness is of the form of Equation 1,

$$T(\log t - c) \qquad (1)$$

where T is the absolute temperature of tempering, t the time and c

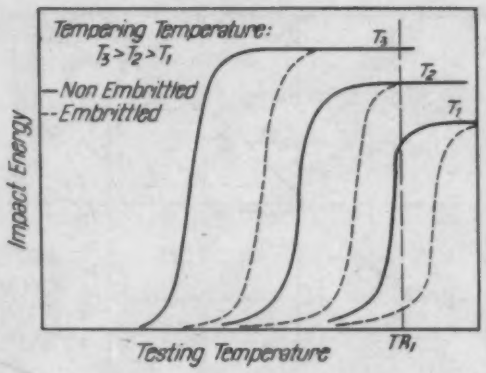


Fig. 21—Effect of Increased Tempering on Impact Properties of a Susceptible Steel.

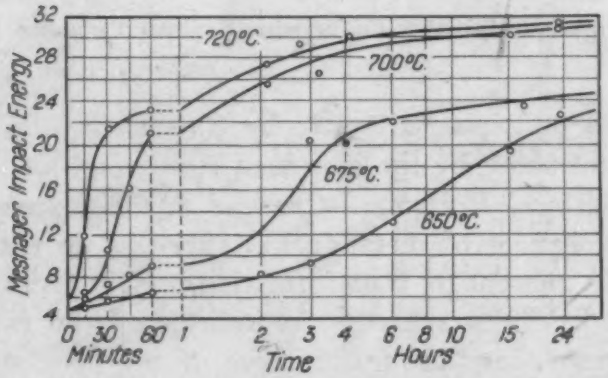


Fig. 22—Effect of Tempering Temperature and Time on Impact Energy After Embrittlement of a Temper Brittle Steel. (After Jolivet and Chouteau.)

a constant which depends upon the steel (not to an important extent). From the data obtained with several other steels (35) a constant was chosen for the steel used by Jolivet and Chouteau. In Fig. 23 the data of Fig. 22 are plotted as a function of the parameter of Equation 1 with a constant of 20 (time in hours). Since the impact energy depends only upon this parameter independent of the time and temperature of tempering, the impact energy for tempering

above 650 degrees Cent. (1200 degrees Fahr.) depends upon hardness and not specifically on time and temperature except as they affect the hardness. This result permits the conclusion that the amount of relative embrittlement is independent of the temperature above about 600 degrees Cent. (1110 degrees Fahr.). An embrittled steel will change its impact properties as it becomes softer even if the amount of transformation remains the same. The effect of repeated tempering and the effect of an increased temperature of tem-

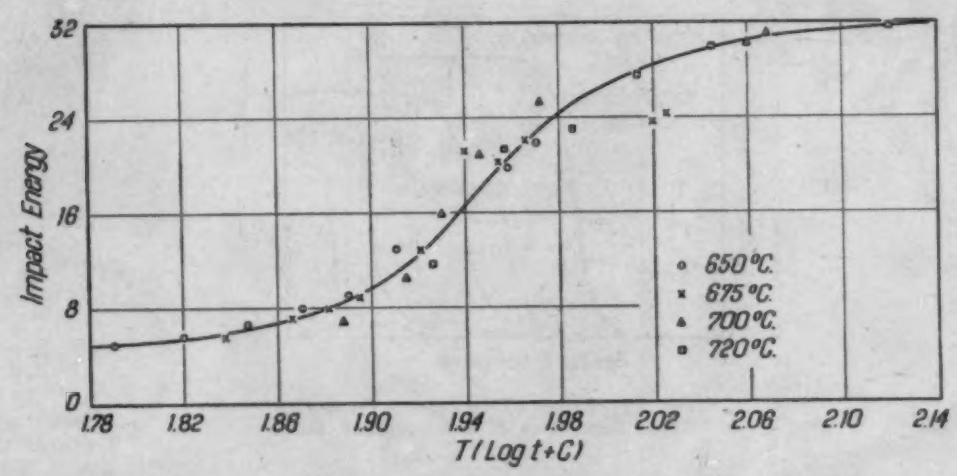


Fig. 23—Variation of Impact Energy with Hardness Parameter for Data of Fig. 24.

pering can both be explained solely by considering the decrease in hardness brought about by the increased tempering.

C. *Metallurgical Variables*—Changes in the susceptibility ratio resulting from changes of metallurgical variables do not necessarily indicate that the amount of transformation arising from the embrittling treatment has been altered. As indicated in an earlier section it could also mean that the impact curve of the nonembrittled steel had been changed. If it is certain that the impact curves of two nonembrittled series of samples (differing with respect to the controlled variable) are the same and there is a change of susceptibility ratio, it must arise from an increased rate of transformation induced by change in nucleation or growth rate or from an actual increase in the total amount of transformation which can occur at each temperature. Further, there can be a change in the effect which a given amount of transformation induces. For these reasons the interpretation of the variation of the susceptibility ratio with metallurgical variables is extremely difficult.

Since the amount of transformation which takes place can only be measured indirectly by impact tests it is almost impossible to evaluate the results of the early experiments which were designed to determine the effects of variables on the transformation. The apparent temper brittleness measured by comparing the impact results of specimens slowly cooled to those rapidly cooled after tempering above 600 degrees Cent. (1110 degrees Fahr.) is useful primarily in determining whether a steel reflects the embrittling to a degree sufficient to cause difficulties in service and does not reflect precisely the amount of transformation which has occurred. The acceptance of susceptibility ratio data, at their face value, leads to such confusion that one is liable to conclude that temper brittleness is such an erratic phenomenon that it depends in no simple way on any variables.

Greaves and Jones (37) stated the problem without elaborating upon it: "The numerical value of the susceptibility ratio of any given steel depends on the conditions of hardening and tempering, the rate of cooling from the tempering temperature, and probably also upon the amount of work put on the steel in forging, rolling, etc., and on the direction in which the tests are made. These conditions should be the same (or should be taken into account) in comparing the susceptibility of different steels". When considering the effects of variables of steel making or composition on the transformation which causes embrittlement it is *not* sufficient to compare the impact properties at one testing temperature of nonembrittled and embrittled specimens. This procedure does not take into account that the mechanical properties of the nonembrittled steels may be markedly different. The sections immediately following are devoted to a discussion of the effects of the major metallurgical variables on the *real* effect of temper brittleness although the data have been obtained in such a manner that even the directions of the effects must in most cases be deduced.

1. *Structural Variables*—In one of the earliest papers concerning temper brittleness Dickenson (8) made the statement that temper brittleness appeared to develop less readily in unhardened steels than in fully hardened ones. Grenet (4) and Greaves and Jones (14) also presented data in which the susceptibility ratios were less for unhardened nickel-chromium steels than for hardened ones. Maurer, Wilms and Kiessler (38), however, conclude in part that if the susceptibility as measured by comparative impact tests at room tempera-

ture is high then slow cooling from the austenitizing temperature reduces the apparent temper brittleness. On the other hand, if the susceptibility is low, slower cooling after austenitizing increases the apparent temper brittleness. This inconsistency of the data as to the effect of prior structure can be rationalized in terms of the earlier discussion. The change in impact properties caused by the change in "as-quenched" structure alters the apparent temper brittleness. Since no quantitative measure of the transformation which causes temper brittleness is available it is not possible to determine whether

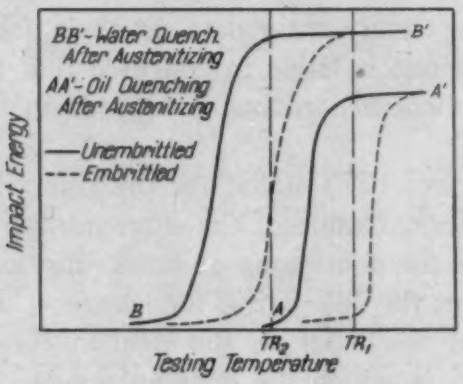


Fig. 24—Effect of Increased Rate of Cooling After Austenitizing on Specimens from Susceptible Steel.

this is the only effect of the change in microstructure. However, if it is assumed that the effect of temper brittleness in shifting the impact curves is independent of the structure of the steels the available data may be rationalized in terms of the discussion of the earlier sections. Consider that the curves AA′ of Fig. 24 are for a steel quenched in oil (from the austenitizing temperature) before tempering, and BB′ for a steel quenched in water (not necessarily to martensite) before tempering. The dotted curves are for the embrittled specimens in both cases and the solid curves are for the nonembrittled specimens. The impact properties of the nonembrittled steels are impaired by slow cooling after austenitizing. Suppose that room temperature is T_{R1}. The curves show that the apparent susceptibility as measured by the ratio of room temperature impact tests is increased by reducing the cooling rate after austenitizing. Now consider the situation if the testing temperature is T_{R2}. In this case a decrease in initial quenching rate and the consequent shift in impact properties will reduce the apparent susceptibility. There are innumerable situations which can be imagined. For example, the

hardness may be different for different specimens, shifting the impact curves and changing the apparent susceptibility even though the amount of the transformation and its effect in shifting the impact curves at each hardness is the same. Both the hardness and the structure may change. It must be concluded that with the available data nothing can be said concerning the effect of structure on the amount of transformation which occurs after a given embrittling treatment or on the magnitude of the effect of a given amount of transformation.

Greaves and Jones (14) as well as Mauer, Wilms and Kiessler (38) have found that increasing the grain size (austenitizing temperature) increases the susceptibility as measured by room temperature impact tests. It seems likely that the observed effect of an increasing susceptibility with increasing austenitizing temperature arises from the effect of the higher quenching temperature (grain size ?) in lowering the impact properties of the nonembrittled specimens and not to an increase in the amount of transformation occurring with a specific embrittling treatment.[11]

2. *Composition Variables*—The interpretation of the voluminous data on the effect of composition suffers from the same problem of separating the effects of the variables. Suppose, for example, that a series of steels is prepared of increasing nickel content. If specimens of the steels are quenched in small sections (large enough for impact specimens) the higher nickel content alloys will be martensitic and the plain carbon steels will consist of mixed structures (pearlite, etc.). If all the specimens are given the same tempering treatment the hardness will vary with changes in the as-quenched condition. Due to these effects and the unavoidable changes in steel "quality" from heat to heat the basic properties of the nonembrittled steels may be widely different. If in changing the composition of a steel other factors such as structure and steel quality vary, the variations will be reflected in the *apparent* temper brittleness as measured by susceptibility ratios. Greaves and Jones (14) state that this unavoidable variation should be taken into account but there is not at present any simple means of so doing. Since in general only room temperature impact tests have been performed it is not possible at present to determine whether, in a given series of experiments designed to determine the effect of changes in steel composition, the properties of each of the steels in the series were identical

[11]It is suggested (page 523) that an increase in grain size might even decrease the rate of transformation.

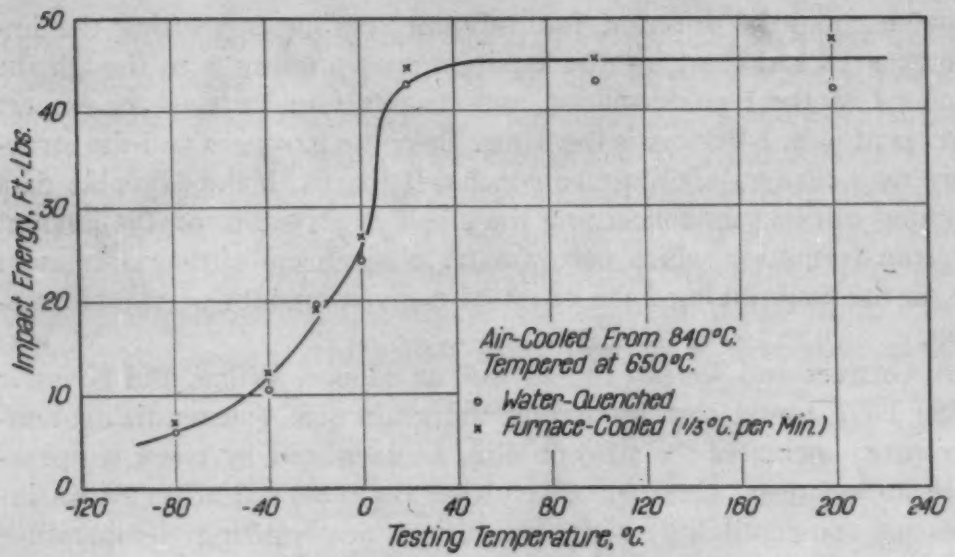


Fig. 25—Temper Brittleness Tests on an S.A.E. 1035 Steel.

in the nonembrittled condition. (It is likely that changes in "as-quenched" structure, hardness, and steel quality were of sufficient magnitude to influence the results.) Because of this difficulty only general trends can be established from the available data.

a. *Plain Carbon Steels*—The situation with regard to plain carbon steels is particularly unsatisfactory for it is seldom that they break in impact tests at room temperature with other than brittle fractures. However, the general trend of the results indicates that plain carbon steel containing less than about 0.60 per cent manganese do not exhibit temper brittleness to an appreciable extent. Tests on a 0.35 per cent plain carbon steel performed by the author are presented in Fig. 25 (Appendix A). Grenet (4)[12] shows data for an essentially plain carbon steel (manganese 0.60 to 0.70 per cent) which appears to exhibit some embrittlement as evidenced by room temperature tests. Greaves and Jones (15) and Nagasawa (33) tested a number of plain carbon steels and found no embrittlement evidenced by room temperature impact tests after furnace cooling from the tempering temperature. Greaves and Jones (15) point out, however, that the impact energy of the plain carbon steels was very low even without the embrittling treatment and hence the results are not too significant. On the basis of these and other data to be presented subsequently it appears that plain carbon steels contain-

[12]It must be remembered, however, that the recent experiments were performed upon aluminum-killed steels while the earlier experiments of Grenet, for example, were not.

ing less than about 0.60 to 0.70 per cent manganese are not susceptible or at least only slightly susceptible to temper brittleness.

b. *Manganese, Chromium and Nickel*—The experiments of the early investigators show only the general trend of the alloying elements in their effect on the amount of transformation which results in temper brittleness. Nickel, chromium, manganese, phosphorus, and to some extent vanadium, all increase the susceptibility to temper brittleness as measured in the standard manner. Molybdenum, and possibly tungsten and columbium, tend to decrease the susceptibility ratio. It is possible, however, that these effects can be accounted for either by a change in the impact properties of the nontransformed steels due to the addition of the alloying elements, to a change in rate of transformation, or to an actual increase in the amount of transformation or in its effect on the impact properties. With the available data it is difficult to separate the influences of these effects. The data of Greaves and Jones (15) indicate that all three elements, manganese, chromium, and nickel, increase the susceptibility ratio for a given manganese content. However, the low-carbon, low-alloy steels definitely did not consist of a martensitic structure on quenching and the susceptibilities of the steels of different alloy contents cannot be compared directly. It appears, however, that these three elements cause the transformation to take place in steels and increase the amount of transformation (as revealed by impact tests) which occurs with a given treatment. More recent work by Greaves (16) indicates that both nickel and manganese increase the susceptibility but that nickel is less effective than manganese.

Recently experiments (Appendix B) were performed on a large number of specimens from castings of various alloy contents each containing about 0.40 per cent molybdenum. Based on these experiments, Fig. 26 is presented in which the reciprocal of the susceptibility ratio (—40 degree Cent. V-notch Charpy Tests) is plotted against the sum of the manganese, chromium and nickel contents. All the specimens were quenched to martensite and if necessary cooled below room temperature before tempering to transform any retained austenite. The level of impact properties at +20 and —40 degrees Cent. were very similar but the impact properties at temperatures lower than —40 degrees Cent. were not determined. Based on experience, however, it is not believed that the alloying elements affect to any considerable extent the impact properties of tempered martensitic steels (of the same hardness). Some differences in be-

havior due to uncontrollable differences in refining and casting practices were no doubt present. The data indicate, however, that all three of the elements increase the transformation for given cooling conditions.

This fact leads to an interesting speculation. Since in designing compositions for obtaining martensite in steel shapes there is frequently some choice as to which alloys will be used to increase the hardenability it is interesting to consider whether the effects of the

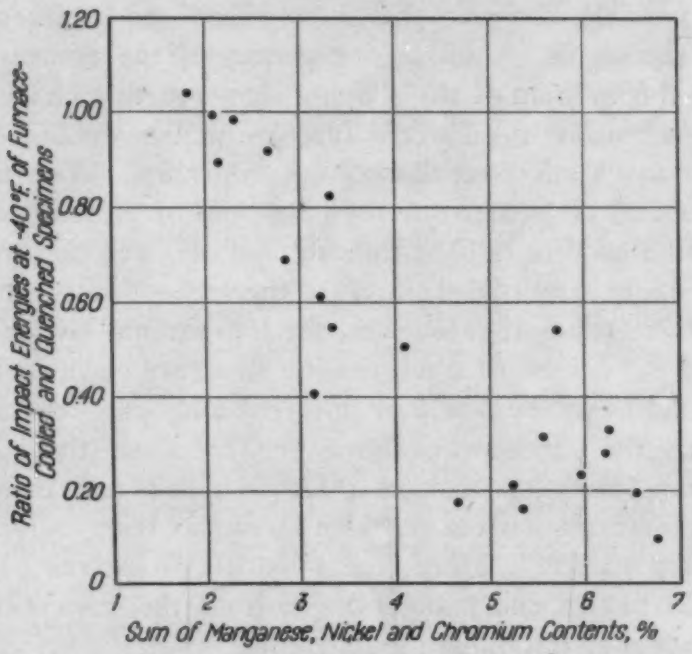


Fig. 26—Effect of Alloying Elements on Susceptibility Toward Temper Brittleness. (See Appendix B.)

chromium, nickel or manganese dictate in any way which of the alloys should be used to increase the hardenability while minimizing the temper brittleness. To put the question in another way: Is there any difference between the effects of alloys in temper brittleness when their different effects on hardenability are taken into account? In Fig. 27 the data of Fig. 26 are plotted against the product (1+XCr) (1+YMn) (1+ZNi) where X, Y, Z are the Grossmann hardenability factors for chromium, manganese and nickel respectively. Thus the ordinate represents a given hardenability arising from the three alloying elements. For the wide variation in alloying elements in the steels there appears to be no significant trend in the data if the alloy combination results in an equal contribution to the

hardenability. This is not to say that susceptibility to temper brittleness and hardenability are necessarily related, but only that if a steel has the same hardenability with respect to the elements chromium, manganese and nickel they will have approximately the same susceptibility toward temper brittleness. The analysis of these data and those of Greaves and Jones (15) indicate that manganese has a greater effect than chromium or nickel.

Other investigators (22), (23), (38) have shown that temper

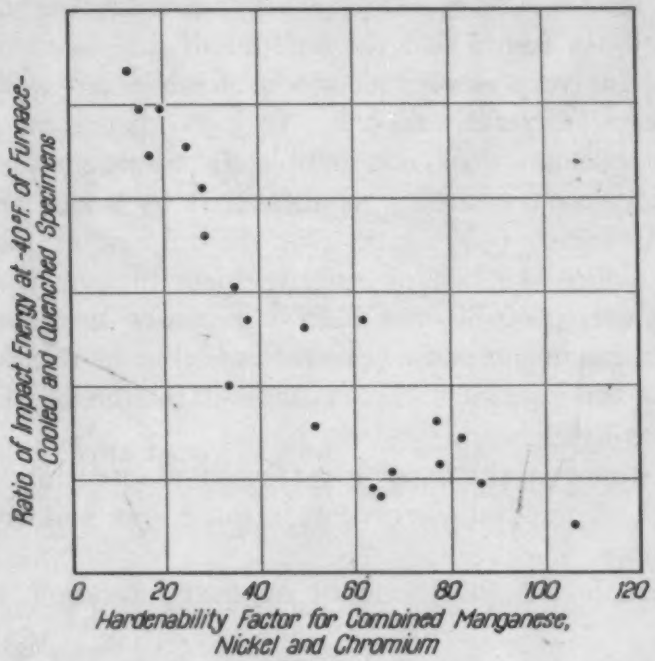


Fig. 27—Susceptibility Ratio of Several Steels for Equal Hardenability as Determined by Manganese, Chromium and Nickel. (See Appendix B.)

brittleness occurs in steels containing manganese, nickel, and chromium and that the amount of transformation tends to increase with increasing amounts of these elements. Lea and Arnold summarized the data on the occurrence of the longitudinal fracture in nickel-chromium steels which is also a manifestation of temper brittleness and indicated that it occurs in steels of high nickel (and chromium) contents.[13] Recent German (39) experiments on aircraft steels containing high manganese or high chromium contents also indicate that such steels are very susceptible to temper brittleness.

[13]It may be of importance that the very high nickel steels did not exhibit the anomalous fracture. It also may be coincidental.

c. *Molybdenum*—Molybdenum (and perhaps tungsten, columbium and titanium[14]) appear to decrease the susceptibility toward temper brittleness. Greaves and Jones (15) found that with nickel-chromium (low manganese) steels the susceptibility ratio could be reduced very nearly to one by the addition of about 0.30 per cent molybdenum. They found also that treatment at 450 degrees Cent. (840 degrees Fahr.) for up to 7 days after tempering at 600 degrees Cent. (1110 degrees Fahr.) did not perceptibly reduce the Izod impact energy (room temperature tests) of nickel-chromium-molybdenum steel. As the susceptibility ratios increased, however, Greaves and Jones found that the amount of molybdenum required increased and for very susceptible steels molybdenum would not reduce the susceptibility ratio to one. In high manganese steels, for example, molybdenum does not completely eliminate the susceptibility (for a given treatment) as measured by room temperature tests. Also Goodrich (40) states that the amount of molybdenum necessary to reduce the rate of embrittlement in long time tests at 450 degrees Cent. (840 degrees Fahr.) increases as the amount of nickel and chromium increases (the susceptibility increases).

There are three possible explanations of this effect of molybdenum on temper brittleness.

1. It so improves the base properties of the nonembrittled steel that the effect of the transformation is made less evident in room temperature tests.

2. It reduces the total amount of transformation which can occur.

3. It reduces the amount of transformation which occurs in a given time (or with a given treatment) by reducing the rate at which the transformation occurs.

The data on the effect of molybdenum do not permit a definite conclusion to be drawn as to which of these three explanations is valid. The data of Lea and Arnold (25) of Fig. 28 in which the impact value is plotted as a function of time of transformation at 500 degrees Cent. (930 degrees Fahr.) (after a 650 degree Cent. temper) indicate that molybdenum simply retards the transformation but does not reduce the total amount which occurs. Other data obtained by these same investigators on similar steels are also consistent with this viewpoint. The data of Maurer, Wilms and Kiessler (38) indicate that molybdenum does act to reduce the apparent suscep-

[14]See the section following.

tibility to temper brittleness as measured by impact tests at room temperature after furnace cooling following tempering and even after long time treatment at 450 degrees Cent. (840 degrees Fahr.). The data indicate, however, that there is an optimum molybdenum content for obtaining this effect. These authors found that with increasing molybdenum content the susceptibility first decreases, then increases. These authors were not careful to obtain complete carbide

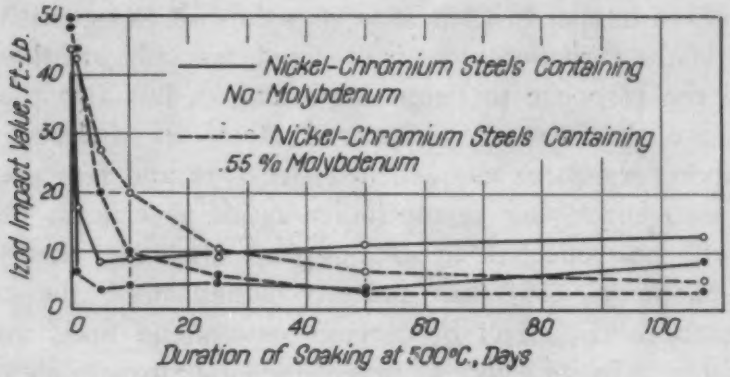


Fig. 28—Effect of Molybdenum in Retarding Temper Brittleness Transformation. (After Lea and Arnold.)

solution and their data may be only a reflection of the change in composition of the austenite. There is no evidence, however, that molybdenum continues to be further effective after about 0.40 per cent has been added.

d. *Other Alloying Elements*—Greaves and Jones state that vanadium increases the susceptibility ratio only slightly but because of the difficulty inherent in the interpretation of all temper brittleness data nothing definite may be said as to the effect of this element.

Tungsten occupies a particularly anomalous situation. Greaves and Jones (15) conclude on the basis of a limited series of experiments that it has no effect on the susceptibility as measured by standard tests. Houdremont and Schrader (22) state that its effect is similar to that of molybdenum (which might be expected). Without further experiments, however, no definite conclusions can be drawn.

A recent Russian paper by Kishkin (41), which does not appear to be readily obtainable in the United States but which has been abstracted in the "Chemical Abstracts", discusses the results of experiments designed to determine the effects of columbium and titanium on temper brittleness. Kishkin found that these elements as well as molybdenum decrease the *apparent* susceptibility as measured by

room temperature tests after the steels had been austenitized at low temperatures. At higher austenitizing temperatures these elements were less effective. More experimental work is certainly necessary to establish the precise effect of these elements.

The data of Bolsover and Barraclough (42) indicate that tin has little if any effect in increasing the susceptibility toward embrittlement (standard susceptibility tests).

e. *Carbon*—The effect of carbon content on the transformation which results in temper brittleness is very difficult to establish experimentally. If the carbon content is changed, not only are the hardenability and the response to tempering changed, but also the impact properties are different for tempered steels of different carbon contents having the same as-quenched structure and tempered hardness. In tests employing susceptibility ratios as criteria these differences will be reflected in changes of apparent susceptibility. Mikhailov-Mikheev (28) has recently summarized the available data on the *apparent* effect of carbon content and finds that there is no consistency in its effect as determined by susceptibility ratios; Greaves and Jones (14) and Lea and Arnold (25) finding that an increase of carbon content increases the (apparent) susceptibility and Cornelius (49) finding that it decreases it.

No definite conclusions can be drawn, at this time, as to the direct effect of carbon on temper brittleness. There is, however, an indirect effect of this element that is worth noting. Frequently, it is desired to obtain a definite tempered hardness in fully quenched steels with some minimum practical tempering time. Because of the effect of carbon content in increasing the hardness for a given tempering treatment, the minimum tempering temperature which may be employed without embrittlement will produce higher hardnesses with higher carbon steels. For this reason, increases in carbon content are sometimes desirable. Other factors such as tendency toward quench cracking must be given consideration however.

f. *Phosphorus*—An impressive amount of data (44), (45), (14) have been collected concerning the effect of phosphorus on the susceptibility ratio as measured by room temperature impact tests. Also one investigator (46) (Bennek) measured the impact energies of ductile and embrittled specimens over a range of testing temperatures. Greaves and Jones (15) summarized much of the data concerning phosphorus and their chart is presented as Fig. 29. Maurer, Wilms and Kiessler (38) performed a number of experiments on

steels of various phosphorus contents but unfortunately they did not indicate the actual impact figures (only ratios), nor the hardness, nor did they perform tests as a function of temperature. From their data (or the data summarized in Fig. 29), it is not possible to determine whether or not the effect of phosphorus arises from a shift of the transition temperature (temperature of brittle fracture) of the nonembrittled specimens or to a real effect of the embrittling treatment. The data of Bennek, however, throw some light on this ques-

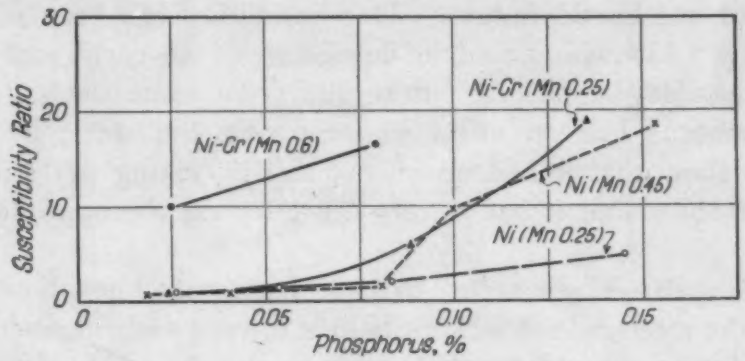


Fig. 29—Relation Between Susceptibility and Phosphorus Content. (After Greaves and Jones.)

tion. This investigator measured the impact energy of high and low phosphorus steels of several alloy contents in both the ductile and embrittled conditions over a range of testing temperature. The data indicate that the relative shift in impact properties is between the high and low phosphorus steels whether in the embrittled or nonembrittled conditions. In other words, the transition temperature is higher for the higher phosphorus content steels in the nonembrittled conditions. It appears from this data of Bennek[15] that the primary effect of the addition of phosphorus is to raise the transition temperature of the unembrittled steel and cause the temper brittleness to become more apparent. It is possible that phosphorus exerts a real influence in increasing the amount or the effect of the transformation but there is no *direct* evidence to support this contention and Bennek's data seem to indicate otherwise. It should be noted that Bennek's data indicate that plain carbon steels of high and low phosphorus content do not exhibit any effect of the embrittling treatment. Bolsover and Barraclough (42) present room temperature impact data for two plain carbon steels containing high and low phosphorus

[15]Bennek did not draw this conclusion.

steels indicating that the high phosphorus steel evidences some temper brittleness under these test conditions. It is not clear, however, that the lower phosphorus steels would not become embrittled at lower testing temperatures. These data on plain carbon steels indicate that if they contain 0.75 per cent manganese they may reflect temper brittleness to a slight extent.

If it is true, as it appears to be, that phosphorus acts primarily to shift the impact curves of the nonembrittled steel, then the effects of phosphorus should be very erratic depending on the exact temperature of the brittle failure. This variability has been noted by Monypenny (43) who stated in discussion of an early paper that: "It was possible, by picking out results from some steels, to show that phosphorus had an effect in improving the steel; by taking others, to show that it had no effect; and by taking out still more they could show that it had a very big effect in the opposite direction".

g. *Oxygen*—There seems to have been only one investigation in which the steels whose susceptibilities toward embrittlement were determined were also analyzed for oxygen. Bischof (23) determined the apparent susceptibility (after water quenching and furnace cooling from 600 degrees Cent.) of a number of heats of nickel-chromium steels and analyzed the steels for phosphorus, oxygen, and nitrogen. No particular trend in the data with nitrogen or phosphorus[16] content could be observed. Bischof in discussion (47) of his paper, however, plotted the percentage decreases in impact energy (room temperature tests) as a function of the oxygen content of the steel. A trend of increasing apparent temper brittleness with increasing oxygen content was observed. The same problems are encountered with the interpretation of these data as with those showing the effects of other alloying elements on the transformation which results in temper brittleness. All of these steels were pearlitic (probably) and the oxygen could have very easily caused a shift of the nonembrittled impact energies to higher temperature causing an increase in the apparent temper brittleness. Until some definite data are obtained it must be concluded that there is no positive evidence of an effect of oxygen on the amount or rate of the transformation which results in temper embrittlement.

h. *Nitrogen*—Apparently there is only one investigation in which the effect of nitrogen has been studied in any way. Griffiths

[16]No major variations of phosphorus were encountered.

(48) introduced nitrogen into several steels by treatment in an ammoniacal atmosphere. He then measured the apparent temper brittleness of the "ordinary" and treated steels. As measured in this manner the nitrogen appears to increase the susceptibility toward embrittlement. The conclusion is the same; the real effect of nitrogen cannot be deduced from the data of Griffiths.

3. *Forming, Casting and Refining Practices*—In the first discussion of temper brittleness Brearley stated a belief which is still

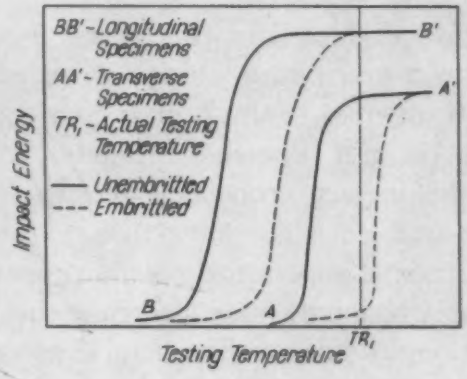


Fig. 30—Effect of Forging Direction on Impact Curves of a Susceptible Steel.

widely held, that the susceptibility to temper brittleness depends upon the method by which the steel is made. As has been pointed out, it is known that such variables will definitely affect the *apparent* temper brittleness. However, there appears to be no evidence to support the contention that the amount or rate of the transformation or the effect of the transformation which results in embrittlement is affected by steel making practice. It is evident from the discussion above that any change in steel making practice which raises the transition temperature will in most cases increase the apparent susceptibility to temper brittleness.

An excellent example of this problem is illustrated by the data Fig. 17 of Jolivet and Chouteau (24). The impact properties of specimens taken transverse and parallel to the forging direction in forged steels are of course quite different. The orientation of the inclusions decreases the impact properties transversely and increases them longitudinally. As a consequence the temperature of brittle fracture is higher for transverse (to the forging direction) impact specimens than for longitudinal. The amount or rate of transforma-

tion giving rise to temper brittleness is probably not[17] different in the two directions but the apparent susceptibility does vary. The schematic curves of Fig. 30 illustrate a possible situation.

No conclusions can be drawn as to the effect of steel making procedures or variables on the amount or rate of formation of the transformation which results in temper brittleness: the *real* susceptibility to temper brittleness.

Mechanism of Temper Brittleness

The information discussed in this paper indicates that temper brittleness is due to a precipitate which forms below about 600 degrees Cent. (1110 degrees Fahr.) and consists of a phase the identity of which is not known definitely. This precipitation affects primarily the impact properties by raising the temperature of brittle fracture and does not alter to any great extent the ordinarily measured room temperature tensile properties. In extreme cases, however, the precipitate may decrease the reduction of area and may produce longitudinal rather than transverse fractures. It appears definite that manganese, chromium and nickel increase the total amount of the phase which can precipitate and the amount which precipitates in a fixed time in the embrittled range. The effect of molybdenum (and possibly tungsten and columbium) is uncertain. For a given time of treatment the effect of the embrittling treatment is less, for molybdenum alloy steels, probably because of reduced rate of precipitation induced by molybdenum (and possibly tungsten and columbium). The effects of other elements such as oxygen, nitrogen, phosphorus, vanadium and titanium are not definitely known. Phosphorus, for example, apparently increases the temper brittleness but it has been shown that this apparent effect may be traced to a rise in the temperature of brittle failure of the nonembrittled steels which shift toward higher temperatures the range over which the temper brittleness precipitate exerts its influence. Carbon, oxygen, nitrogen, etc., may exert similar influences but definite conclusions as to the effects of these elements do not appear to be warranted.

Steel making practice has been considered to have a marked influence on temper brittleness. It is definite that this variable affects temper brittleness as measured under specific testing conditions but

[17] It is possible that the transformation will occur preferentially in certain regions (bands) due to segregations.

it is not known whether these effects are due to changes in the properties of the nonembrittled steels or to a real change in the amount of precipitate which occurs under a given embrittling condition. Until specific data on the shift of the transition temperatures induced by a specific precipitation treatment are available for steels made by different processes, the conclusion that the *real* effect of temper brittleness varies from heat to heat or among steels made by various processes is probably unwarranted.

A complete explanation of the phenomenon of temper brittleness actually consists of two distinct parts. First, the general features of the transformation which results in the embrittlement must be understood. The reversibility of the process must be considered as well as the fact that the embrittling phase appears to be located almost completely at grain boundaries. The responsible constituent must be identified and the effects of all elements on the transformation rationalized. Second, a mechanism which consistently accounts for the effect of this grain boundary precipitate on the mechanical properties must be formulated. Because more is known regarding the second phase of this problem it will be discussed first.

A. *The Effect of the Precipitation*—The precipitate affects primarily the notched-bar properties of steels in the manner described in the earlier sections of this report, without affecting to any great extent the normal tensile test results. In extreme cases of embrittlement, however, the fracture stress and the reduction of area may be reduced and in some cases the fractures of the tensile specimens are longitudinal rather than transverse. In order to understand these effects of temper brittleness it is necessary to understand by what mechanism a change in structure affects the impact properties without appreciably (or consistently) affecting the normal tensile test results. A recent discussion (50) of the significance of the notched-bar impact test includes a rather detailed analysis of this problem.

The stress-strain curve obtained from an ordinary tensile test may be considered to be the tests of an infinite number of specimens of different histories rather than a test of a single specimen. Each of these specimens has an identical prior history with regard to manufacturing practice and heat treatment but differs from every other specimen in that it has undergone a different amount of tensile strain. Thus the tensile stress-strain curve may be considered to be a locus of the stresses required for plastic flow of an infinite

number of specimens under the particular conditions of the tensile test. At the termination of the tensile test that specimen fractures. It is effective to consider that had not the other specimens undergone plastic deformation they would have fractured at a value of the tensile stress higher than the values given by the locus of flow stresses. This viewpoint requires that there be a locus of fracture

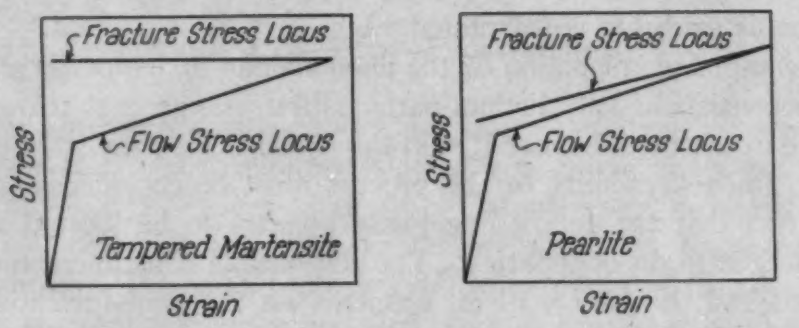


Fig. 31—Schematic Tensile Stress-Strain Curves of Steels Having Typical Structures.

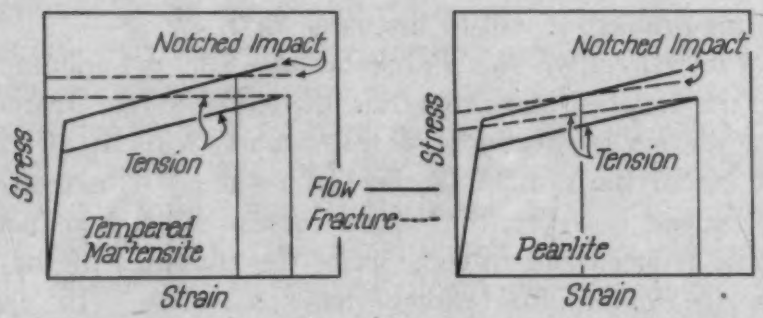


Fig. 32—Schematic Stress-Strain Curves of Typical Structures (Notched-Bar Impact).

stresses which intersects the locus of flow stresses for the specimen which fractures. Fig. 31 schematically illustrates fracture stress and flow stress loci for a tempered martensitic and pearlitic steel (no temper brittleness).

The loci of flow stresses (the flow stress-strain curves) of the two steels may be identical while the loci of fracture stresses (the fracture stress-strain curves) differ widely. No intimation of the shape of the fracture stress-strain curve is revealed by normal room temperature tensile tests. If, however, the two steels are subjected to notched-bar impact tests great differences in behavior may be revealed. An example of the behavior of the two types of steels when *notched* and broken is illustrated in Fig. 32. The energy (per unit

volume) required for fracture of the two steels will be very different. The shift in the flow stress curve is caused primarily by the transverse constraint[18] at the base of the notch and in part by the increase in rate of strain induced by the stress concentration at the base of the notch (and the increased rate of loading if it is a notched-bar *impact* test). The fracture stress curves appear to be unaffected

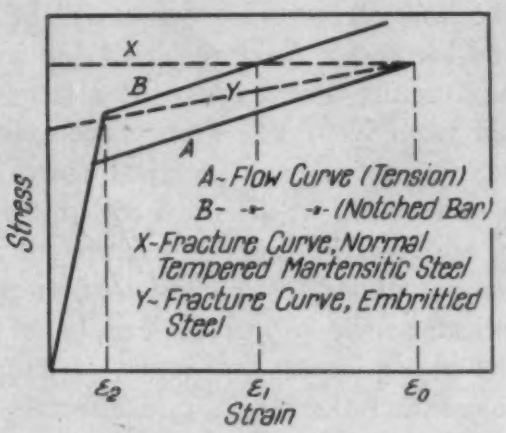


Fig. 33—Mechanism of Effect of Temper Brittleness on Stress-Strain Relations.

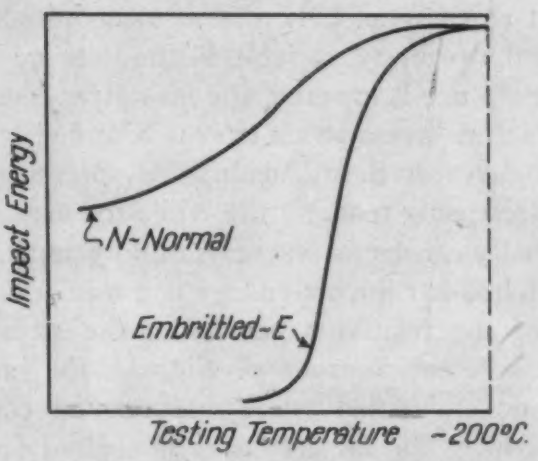


Fig. 34—Effect of Temper Brittleness on Impact Curves of Tempered Martensite Steels.

by the transverse constraint and vary only because of the change in the rate of strain. The effect of strain rate on the fracture stress curves also appears to be less than on the flow stress curves.

The only consistent mechanism of the observed effects of the transformation which occurs in the range of temperature from 250

[18]As pointed out in an earlier report (51), the stress gradient at the base of the notch may in itself raise the flow stress curve.

to 600 degrees Cent. (480 to 1110 degrees Fahr.) (in susceptible steels) is that the fracture stress curve is changed from the type given in Fig. 31a to that illustrated in Fig. 31b. Fig. 33 illustrates this interpretation of the effect of the transformation resulting in temper brittleness. A, is the flow stress-strain curve for both the embrittled and nonembrittled specimens at the strain rate which is encountered in the notched-bar impact test. Because of the transverse constraint the flow stress-strain curve will be shifted from A to B in the notched-bar test. As a result of the temper brittleness transformation the fracture stress curve of a tempered martensitic steel will be shifted from X to Y. Thus there exists the following situation: The two specimens have identical normal tensile properties but the strain to fracture in the notched-bar test of the embrittled specimen is ε_2 and for the untreated tempered martensitic steel it is ε_1. This analysis illustrates a consistent mechanism whereby steels having identical tensile properties can have widely different notched-bar or notched-bar impact properties. It is important at this juncture to emphasize that the strain rate effect is relatively small since this variable affects both the flow and fracture curves to a slight extent but modifies their separation even less. It was Philpot (7) who pointed out as early as 1917 that it was the *notch* and not the *impact* which is the primary variable in the notched-bar impact test.

As the temperature is lowered, the flow stress curve B rises with respect to the fracture stress-strain curves X and Y. The area under the stress-strain curve of the nonembrittled specimen decreases only gradually with decreasing temperature while the area under the curve for the embrittled steel decreases very rapidly until brittle fracture occurs. The notched-bar impact energy is a measure (averaged over the specimen) of the relative areas under the stress-strain curves. Based upon the schematic curves of Fig. 33, the impact energy of embrittled and nonembrittled specimens can be constructed schematically (Fig. 34). The impact curve is shifted from curve N to curve E because of the shift in fracture strength curve from X to Y. These schematic curves are similar to the actual impact curves of a susceptible steel having a tempered martensitic structure (probably) and in the normal and embrittled conditions shown in Fig. 1.

Quenched steels, however, do not always consist entirely of martensitic structures and the fracture strength curves of such steels after tempering are not always of the type given by X of Fig. 33. Pearlitic steels, for example, have stress-strain curves as illustrated

schematically in Fig. 35. (These curves are drawn for the strain rate encountered at the base of the notch in the notched-bar impact test so as to make the figure simpler). The transverse constraint raises the flow curve from A to B and, as the temperature is lowered,

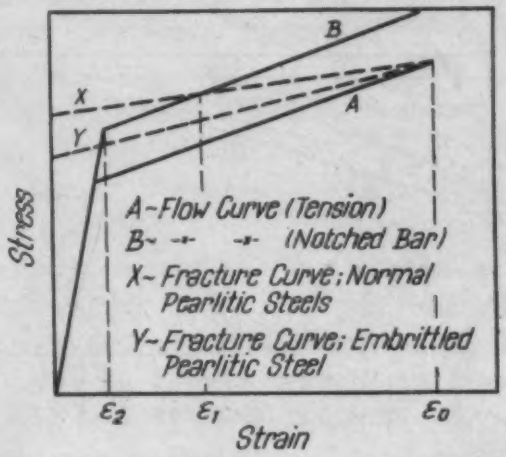


Fig. 35—Effect of Temper Brittleness on Stress-Strain Relations of a Pearlitic Steel.

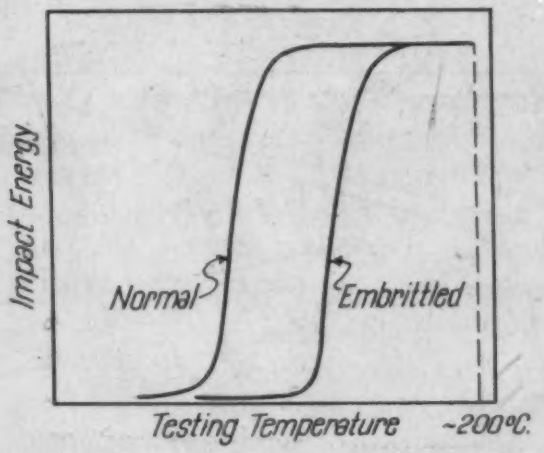


Fig. 36—Effect of Temper Brittleness Transformation on Impact Curves of Pearlitic Steels.

the impact energy curves will have the form given in Fig. 36. This shifting of the temperature of brittle failure agrees with the data (Fig. 2).

Actually it is not believed that the fracture stress-strain curves vary precisely according to Figs. 33 and 35, even though these diagrams suffice to explain the more puzzling features of the temper brittleness phenomenon. The schematic fracture stress-strain curves

of Fig. 37 are more likely and have a possible physical interpretation. This shifting of fracture curves serves to explain the shift in the temperature of brittle failure while still indicating that in severe cases of embrittlement the reduction of area in the normal tensile test may be reduced (as is observed). This type of shifting of the frac-

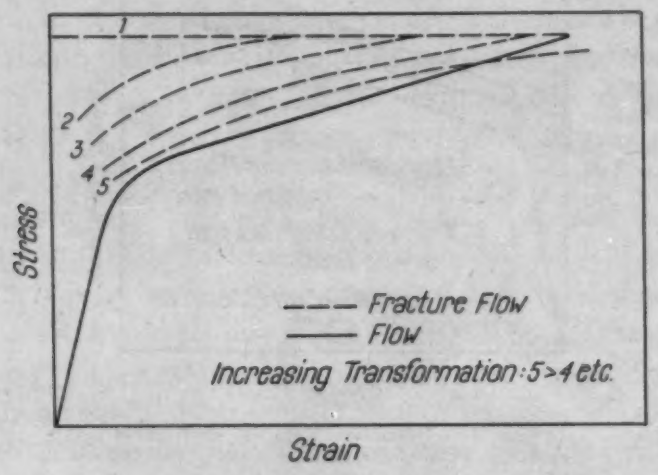


Fig. 37—Postulated Progressive Effect of Temper Brittleness on Fracture Curves.

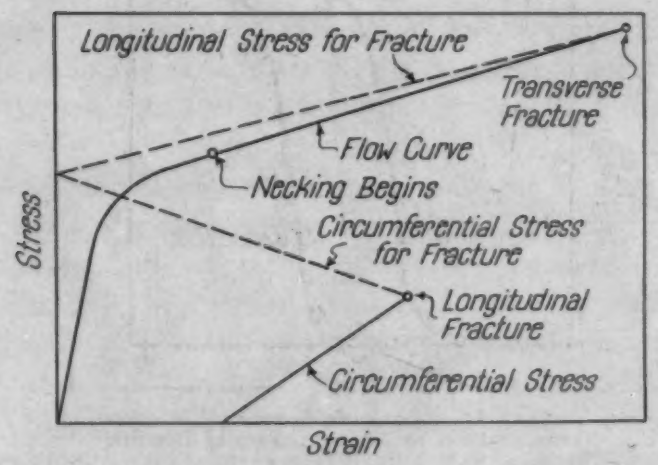


Fig. 38—Schematic Diagram Illustrating Mechanism of Longitudinal Fracture of Temper Brittle Steels.

ture curves may be applied to pearlitic steels as well as to the tempered martensitic. This explanation of the effect of temper brittleness also accounts for the effect of temper brittleness observed at high testing temperatures (notched-bar impact tests).

The sharp rise in the fracture stress may be a result of the de-

struction of the continuity of a grain boundary precipitate as the deformation proceeds, until at large strains the grain boundary no longer exerts its influence on the stress required for fracture. This reasoning assumes that the grain boundary precipitation lowers the fracture stress of the undeformed metal. This lowering is consistent with the viewpoint previously expressed by Zener and Hollomon (51) concerning the factors governing fracture.

The puzzling longitudinal fracture observed with embrittled steels may be understood when it is realized that, if the fracture strength of a temper brittle steel is strongly dependent on strain, the tensile stress required to fracture the specimen in a plane perpendicular to the extension increases and the stress required to break the specimen transversely decreases. (Consider the analogous case of a hot-rolled plate—the tensile stress required to fracture a specimen taken parallel to the direction of rolling increases while the fracture strength of a transverse specimen decreases.) Furthermore, after necking commences the stresses in the necked section are no longer uniform. At the neck there are radial and circumferential tensions as well as a longitudinal tensile stress. Bridgman (52) has recently analyzed the stress system at the neck of a tensile specimen in detail. His analysis indicates that at a strain of one the radial and circumferential tensions become equal to about 40 per cent of the longitudinal stress (at the axis of the specimen). Consider now the schematic diagram of Fig. 38. The longitudinal stress required to fracture the specimen (transversely) increases with strain; the circumferential stress required to fracture the specimen (longitudinally) decreases with increasing strain. As the deformation continues after "necking" the circumferential (and radial) stress increases. If the slope of the fracture strength curve is sufficiently great fracture will occur longitudinally under the action of the circumferential stress before the normal transverse tensile fracture can occur. This type of fracture should also occur in very ductile pearlitic steels for as has been pointed out previously the fracture strength of such steels depend markedly on strain. This argument serves in a sense to confirm the analysis of the effect of temper brittleness on the mechanical properties. Further, it is a consistent mechanism for the explanation of the longitudinal fracture and to the frequently observed "star" fracture.

B. *The Nature of the Precipitation*—Many speculations have been made concerning the transformation which results in temper

brittleness. The various hypotheses have been succinctly summarized by Jolivet and Chouteau (24)[19]:

"Diverse hypotheses have been proposed which account more or less perfectly for the observed phenomena.

(a) Hypothesis of an unknown transformation of existing phases (ferrite or carbides).

"There have been suggested the existence of a transformation below 700 degrees Cent. [Le Chatelier (53)], of an allotropic modification of the iron [Jeffries (11)], of a variation in the molecular arrangement of the atoms [Hatfield (6)], of a modification of the carbides [Maurer and Hohage (54)];

(b) Hypotheses of a retardation in the known transformations.

"The brittleness has been attributed to the decomposition of retained austenite into martensite [Esser and Eilander (55)];

(c) Hypothesis of the precipitation of a new phase.

"This hypothesis has been formulated many times. In particular, there has been suggested that the intergranular precipitation of cementite or carbide is responsible: [Rogers (30)], [Grenet (56)], [Andrew and Dickie (17)], [Honda and Yamada (57)], [Nagasawa (33)]; a precipitate of chromium oxide [Greaves and Jones (14)]; a precipitate of nitrides [Griffiths (48)]; of phosphides [Feszenko and Czopiwski (58)], [Bennek (46)].[20]

The complexity of the observed phenomena and the difficulties inherent in the study of them explain the fact that a great number of hypotheses have been formulated.

"In summation, it does not appear possible to choose between two of these theories:

1. *The theory of the precipitation,* even though it is not possible to determine precisely the phases which precipitate.

"Certain phenomena, such as aging or structural hardening, are accompanied by a marked increase in brittleness and a visible precipitation.

"In the present case there is no direct evidence of a precipitate. It is only by analogy that the decrease in toughness can be attributed to a precipitation, whose nature is connected with elements which affect the intensity of the temper brittleness.

2. *The theory of the modification of special carbides*—"This

[19]Translated freely by the author and Mrs. J. H. Hollomon.

[20]Several American metallurgists have suggested that temper brittleness arises from a simple precipitation. Among these are French (59), and Merica (60). (See also Newell's (61) discussion to paper by Wilten who suggests a grain boundary precipitate on the basis of intergranular cracking of embrittled chromium steels.)

hypothesis formulated by Maurer and Hohage has been recently revived by Bischof (47) and by Maurer, Wilms and Kiessler (38). In this case the brittleness would be connected with the existence of a carbide containing alloying elements and susceptible to decomposi-

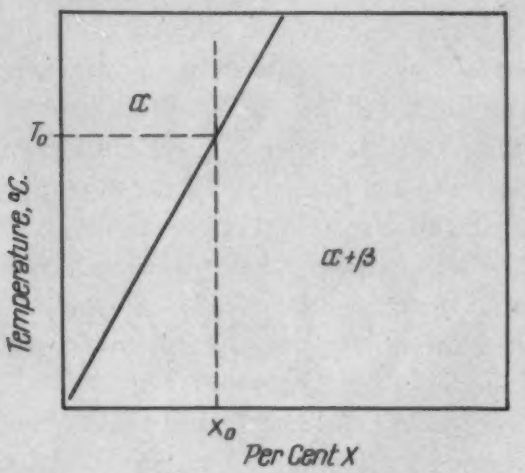


Fig. 39—Phase Diagram for Precipitation from Solid Solutions.

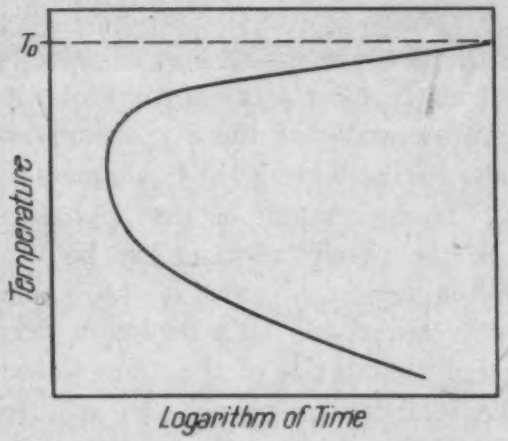


Fig. 40—Relation Between Time and Temperature for Initial Precipitation from Solid Solution.

tion to a carbide less rich in alloying elements with the liberation of chromium and manganese. These elements will go progressively (at higher temperatures) in solution resulting in the disappearance of the brittleness."

It appears that all the available data can be interpreted in terms of a simple precipitation from solid solution.[21] (See above). Sup-

[21]The mechanism of precipitation has been summarized in part by Mehl and Jetter (62).

pose, for example, that the precipitate is a phase (β) composed of iron and another element X. The variation of the solubility of X in alpha iron will have the form given in Fig. 39. Assume an alloy containing X_o per cent of the element X. If a metal of this composition is heated above the temperature T_o all the X will tend to go into solution in the phase α. As the alloy is cooled below the temperature T_o the phase β will precipitate at a rate determined by the variation of the solubility of X in α and by the rate of nucleation and growth of the β phase (which depends on the temperature of undercooling). Just below the temperature T_o the rate of transformation must be zero because, among other reasons, the rate of nucleation is zero. At very low temperatures the rate of diffusion is very small and the rate of transformation must also be small. Thus the time required for a given amount of transformation will have the form given by Fig. 40.

It should be pointed out, however, that this curve represents the time required to produce a given amount of β in terms of per cent of total alloy and not in terms of amount which can precipitate (Fig. 39).

Qualitative reasoning indicates that in all cases of precipitation from solid solution the first precipitate will form at the grain boundaries, occurring first at the most acute corners of the grains[22]. The nuclei will form preferentially at these positions since the energy required to form new surfaces is less at boundaries (and still less at corners of grains). If the amount of the β phase which can precipitate is large then the precipitation at the boundaries will cease due to depletion of the region adjacent to the boundary in X soon after the boundary is covered. This depletion decreases the rate of diffusion of X from the interior of the grain to the boundary, and precipitation will be retarded or new nuclei will form within the grain permitting further transformation. An example of initial grain boundary precipitation, depletion of the metal adjacent to the boundary and later random precipitation is illustrated for a quenched and tempered Muntz metal in Fig. 41.

[22]Nuclei will form at the locations where the energy required for their formation is minimum. Thus the nuclei will form at boundaries for there the energy required for their formation is less than if the nuclei formed in the interior of the grain. This same argument holds for the acute corners of the grains. A question arises, however, as to whether in the case of steel the precipitate will form at ferrite-ferrite, ferrite-cementite or on the old austenitic boundaries. All other things being equal the precipitate will form at those boundaries across which the binding is least. Because of the immobility of the inclusions at the prior austenitic boundaries it would be expected that the least surface energy would be required for nucleation at these boundaries. This argument, however, does not exclude precipitate at ferrite boundaries or even random precipitation; it only requires that the precipitation should occur ***preferentially*** at the austenitic boundaries.

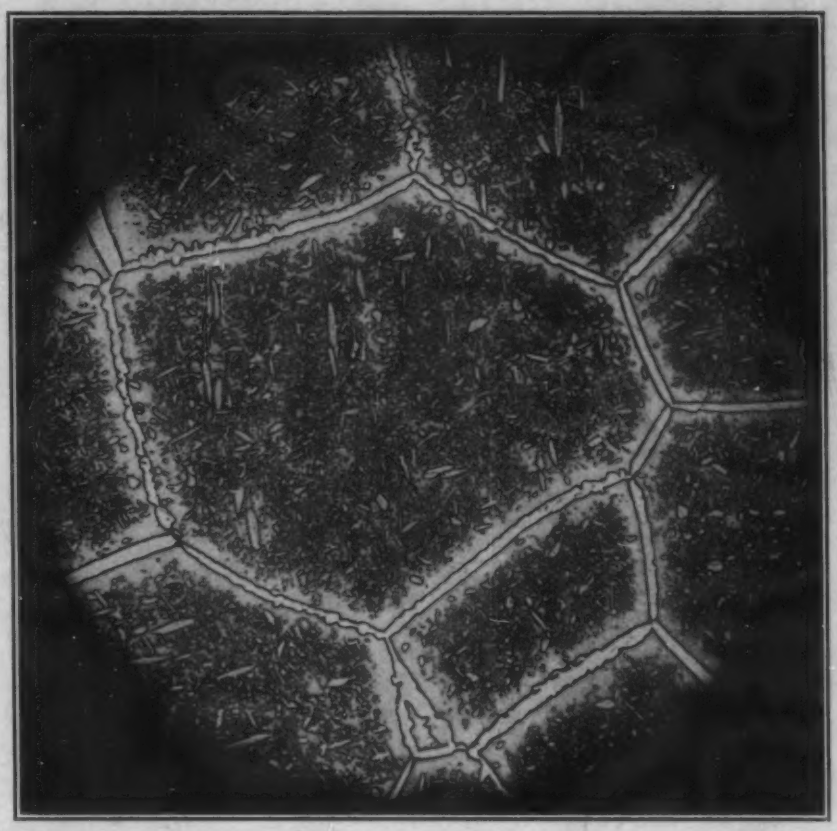


Fig. 41—Random and Grain Boundary Precipitate in Muntz Metal.

If, however, the total amount of the constituent X which can precipitate is small, then the precipitate will occur primarily at the boundary. It can also be argued that the rate of transformation in terms of the fractional amount occurring at each temperature decreases continuously as the temperature is lowered. While the rate of precipitation of β first increases then decreases as the temperature is lowered, the amount of which can precipitate is zero at T_o and increases continuously as the temperature of undercooling decreases. A semi-quantitative[23] argument indicates that the logarithm of the

[23]If the process of the formation of the grain boundary nucleus is the formation of thin film of precipitate at the boundary followed by the growth of this precipitate by diffusion from the interior of the grain the time required to complete the precipitation should vary according to the following relation:

$$t \sim \frac{R^2}{D}$$

where R is the radius of the austenite grain, and D the diffusion constant. Thus the time required to complete the transformation should be related to the temperature as:

$$t \sim e^{Q/RT}$$

where Q is the diffusion constant for the element X, R the gas constant, and T the absolute temperature. This argument would indicate that the amount of precipitate with a given (moderate) embrittling treatment should decrease with increasing austenite grain size. The evidence on this point is not conclusive. (See page 501.)

time required to produce 100 per cent transformation at each temperature should vary linearly with the reciprocal of the absolute temperature.

The data presented in Figs. 7, 8, and 9, in the earlier sections of this report illustrate exactly these relations between the time temperature and transformation for the transformation which results

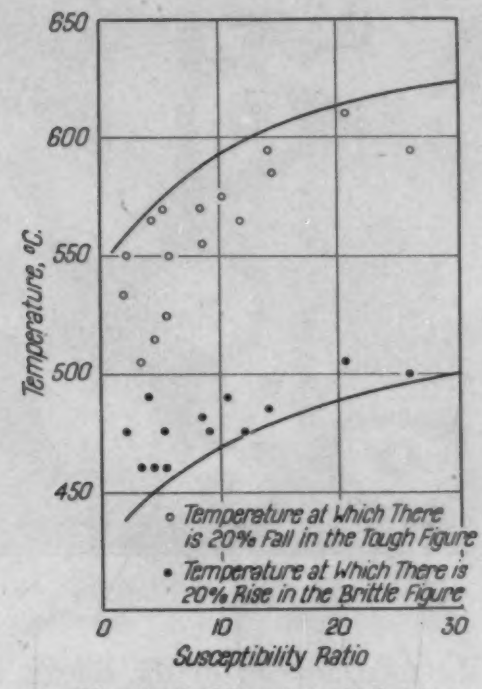


Fig. 42—Position of the Embrittling Range. (After Greaves and Jones.)

in temper brittleness. The arguments used to interpret the effects of the transformation indicate that the precipitate probably occurs primarily at the grain boundaries. It is also indicated that after the equilibrium precipitation occurs holding for longer times tends to slowly improve the impact figure. This improvement can be accounted for by gradual spheroidization of the initial precipitate. A grain boundary (discontinuous) precipitate would not be expected to affect the hardness, electrical resistance or other physical properties (62). It does, however, have a marked influence on the fracture stress.

The available data indicate, however, that the temperature at which the solubility of X in alpha iron is exceeded does not vary to any great extent from steel to steel. Greaves and Jones (14), for example, summarize part of this information in Fig. 42. This

figure shows that the temperature required to produce a given fall in impact value (equilibrium) depends to some extent on the severity of the temper brittleness.

Actually, however, the less severe is the possible embrittlement the lower is the temperature at which a given fall in impact energy occurs. Thus it must be concluded that at least for the steels of Fig.

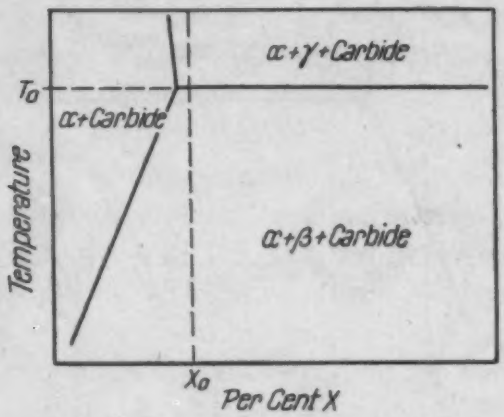


Fig. 43—Suggested Phase Diagram for Temper Brittleness Transformation.

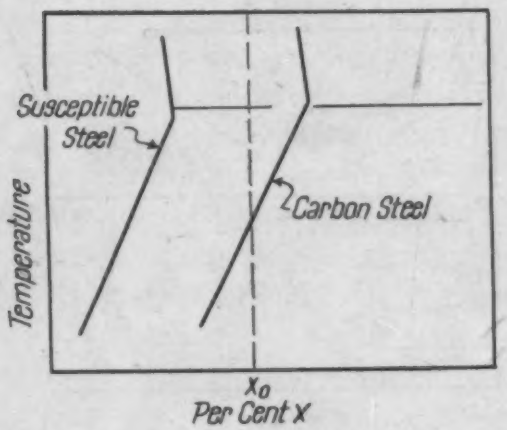


Fig. 44—Possible Mechanism of Increasing Susceptibility (Solubility Change).

42 the temperature at which the steel becomes saturated with respect to the element responsible for temper brittleness is very nearly constant and very nearly 600 degrees Cent. (1110 degrees Fahr.). Heating for sufficient time above this temperature will cause the restoration of the impact properties. Whether or not the solution of the precipitate which causes temper brittleness is complete above 600 degrees Cent. (1110 degrees Fahr.) is not known. The apparent

constancy of the temperature T_o with changes in steel composition is surprising. It requires that the solubility of the element X in the alpha iron is slightly different from that shown in Fig. 39. Since the percentage of the element responsible for temper brittleness must vary from steel to steel and the exact position of the solubility line must vary with alloy content it is not possible that the transformation

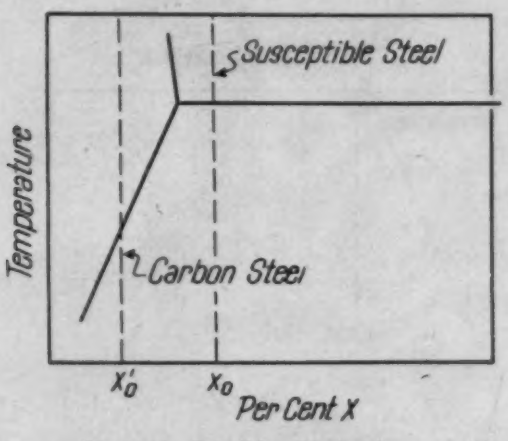


Fig. 45—Possible Mechanism of Increasing Susceptibility (Composition Change).

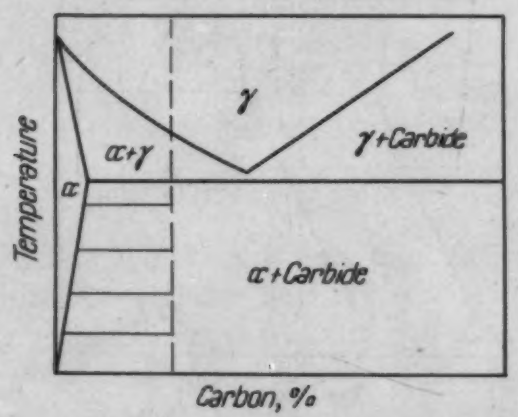


Fig. 46—Iron-Carbon Phase Diagram.

be as simple as that illustrated in Fig. 39. If, however, the solubility of element X in a susceptible steel had the form illustrated in Fig. 43, the approximate constancy of the temperature T_o for a wide variety of susceptible steels is explained. Above the temperature T_o there will be formed a new phase but the amount formed can be extremely small. The effect of the alloying elements manganese, nickel and chromium could be to shift the solubility line to the left as in Fig.

44, or to increase the amount of element X (X_o to X_o') present as in Fig. 45.

In either of these two cases the amount of the constituent which could precipitate is very small in plain carbon steels and it must occur at very low temperatures. Only within a very narrow range of compositions would the critical temperature (T_o) depend upon the alloy content. In order to establish just exactly the effects of the alloying elements it is necessary to discover the element responsible for temper brittleness. At the present time there are no direct experiments available by which its identity may be established. The speculations which have been made are summarized earlier. It is possible, however, to eliminate certain of the possibilities.

One of the most frequently suggested explanations of temper brittleness is that it is connected with the precipitation of a carbide or with a change in type or form of a carbide phase. This suggestion has found such wide acceptance because, superficially at least, the known decrease in solubility of carbon in alpha iron gives a satisfactory mechanism for a precipitation from solid solution. There are several features of the transformation which gives rise to temper brittleness which the precipitation of a simple carbide does not appear to explain. For example, consider the case of two steels, one susceptible to temper brittleness and one not susceptible. If specimens of both of these steels are quenched from the austenitizing temperature and tempered in the embrittling range, the susceptible steel will be embrittled and the nonsusceptible steel will not. Carbides precipitate during the tempering of both steels. Furthermore, the precipitate in embrittled susceptible steels may be made to virtually disappear on heating between 600 and 700 degrees Cent. (1110 and 1290 degrees Fahr.), and precipitation only occurs below 600 degrees Cent. (1110 degrees Fahr.). According to Fig. 46, however, precipitation of carbide from alpha iron should occur at all temperatures below the eutectoidal temperature. It is also known that carbides precipitate upon slow cooling from 700 degrees Cent. (1290 degrees Fahr.), in low carbon steels. If rapidly cooled the steel can be supersaturated with respect to carbon which will precipitate even upon aging at room temperature. During this aging the temperature of brittle failure increases but probably no more than would be expected from the increase in hardness. It appears therefore from these arguments that the precipitate which is responsible for temper brittleness is not a carbide.

The diffusion constant Q (given by the slope of the straight line of Fig. 9) is, however, of the order of magnitude of 14,000 calories per gram molecule. The diffusion constants of carbon and nitrogen in austenite are approximately 30,000 calories per gram molecule and the constants for the other (substitutional) elements are larger. Because of the looser packing of the B.C.C. lattice of the alpha iron it might be expected that the diffusion constants of the interstitial elements nitrogen and carbon would be reduced while those of the other elements would remain about the same. It is

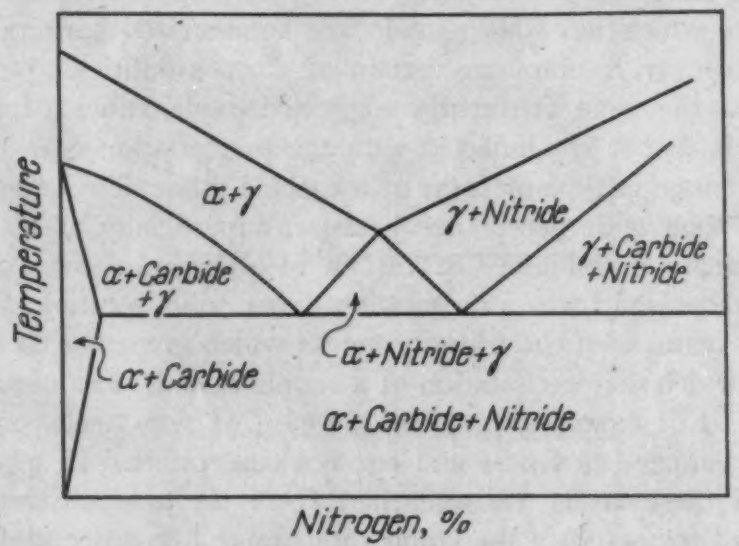


Fig. 47—Schematic Constant Carbon Section of Iron-Carbon-Nitrogen Ternary System for Small Nitrogen Contents.

possible therefore that the element which is diffusing has a low heat of diffusion (relatively high rate of diffusion) and that the element is either carbon or nitrogen. Furthermore, the eutectoidal temperature of the iron-nitrogen system is about 600 degrees Cent. (1110 degrees Fahr.). The ternary eutectoid would be expected to be somewhat lower, how much lower is not known. The iron-nitrogen diagram for a steel containing about 0.30 per cent carbon is presented schematically as Fig. 47. The exact solubility of nitrogen in alpha iron is not known for a wide divergence exists between the results of the various investigators (63). It is believed, however, that the elements manganese, nickel and chromium will act to decrease the solubility of nitrogen in alpha iron and hence increase the amount which can precipitate. This argument offers a possible mechanism for the precipitation. It must be remembered, however, that

nitrogen does precipitate in relatively low carbon-low alloy steels and is at least partially responsible for the inhomogeneous yielding phenomenon (64) and for strain aging. The phase diagram of Fig. 45 would account for this phenomenon in low alloy steels. This diagram would indicate that the quench aging arising from the precipitation of nitride could be made to disappear by a relatively low temperature treatment followed by a quench. It is not known whether such experiments have been performed.

If the precipitation of a nitride is the cause of the temper brittleness it is expected that the amount of precipitation should vary somewhat from heat to heat and process to process as well as with composition. As pointed out earlier in this report there is, however, no evidence at present to warrant this conclusion (not that it may not be true).

Experiments should be performed in which nitrogen is eliminated from susceptible steels in order to test the hypotheses that the precipitate of a nitride is responsible. Furthermore, this survey illustrates a great dearth of reliable information as to the effects of nitrogen on phase equilibria in steels. Because of possible tremendous effects of nitrogen on properties of steels, information as to the effects of nitrogen is very necessary.

The acceptance of a simple precipitation mechanism to account for temper brittleness appears to be inescapable. The suggestion that nitrogen is the responsible element can only be tentative until such time as reliable, quantitative information is obtained on the transformation which is the cause of temper brittleness. An investigation of nitrogen as a possible cause seems to be warranted on the basis of the available information.

Summary

An understanding of the problem of temper brittleness is vitally important to the development of steels having the optimum combinations of mechanical properties. The difficulties caused by this phenomenon became widely recognized during the development of armor and gun steels for World War I. Unfortunately, there is little information concerning temper brittleness in addition to that obtained during this early period. However, much was learned during the years between the two world wars concerning the properties of steels and the transformations which occur during their

heat treatments. On the basis of these new ideas many of the conclusions resulting from the earlier studies of temper brittleness are revised in this paper.

It appears that temper brittleness is caused by a precipitation from alpha iron which occurs preferentially at the (prior austenitic) grain boundaries. The assumption that the precipitate is a phase consisting of iron and an element which supersaturates alpha iron at low temperatures is consistent with the known facts. Above about 600 degrees Cent. (1110 degrees Fahr.) the solubility of the responsible element in steels susceptible to temper brittleness is such that the embrittling precipitate redissolves. Precipitation will occur at all temperatures below about 600 degrees Cent. (1110 degrees Fahr.), but at a rate which is dependent upon temperature. At temperatures just below the solubility limit ($\sim$ 600 degrees Cent.), the precipitation takes place only very slowly, for both the rate of nucleation and the rate of growth are small. At low temperatures the precipitation occurs slowly because the rate of growth is small since the rate of diffusion is small. At some intermediate temperature (about 500 degrees Cent.) the precipitation occurs at a maximum rate.

Thus the time-temperature-transformation curve for the precipitation has the form of a "C" similar to the curve for the decomposition of austenite into pearlite. Even though the rate of growth in terms of absolute amounts of precipitation first increases then decreases as the temperature is lowered, the time required for the completion of the precipitation increases continuously with decreasing temperature.

The precipitate which forms below 600 degrees Cent. (1110 degrees Fahr.) can occur either on tempering after austenitizing and quenching, on cooling from the temper, or on reheating after quenching from the temper. The precipitate manifests itself primarily in an increase of the temperature of brittle failure of notched-bar impact specimens. In room temperature tensile tests the fracture stress and reduction of area are reduced only with severely embrittled steels. The effect of the precipitate can be accounted for only in terms of a shift in the fracture stress curves with increasing precipitation. The mechanism of the effect of the precipitate on the notched-bar test results is discussed in detail. An important consequence of this discussion is the conclusion that the apparent susceptibility as determined by the ratio of the impact energies of an arbitrarily embrittled

specimen and a nonembrittled specimen is very deceptive. For example, the susceptibility of a steel as determined in the above manner will vary with the "as-quenched" structure. Since almost all of the data regarding the effect of variables on the temper brittleness transformation are in terms of the variation of susceptibility ratios, no definite conclusions may be drawn with regard to the effects of these variables. It is necessary to design a test which will permit quantitative measurement of the effect of variables on the rate and extent of the precipitate, and the relation between the amount of precipitate and its effect on the impact properties. As yet no such test is available.

The inadequacy of the data regarding the effect of variables on the temper brittleness transformation permits only certain general inferences to be made. Contrary to popular belief, no conclusions are justified as to the effect of steel making practice on the extent or rate of the transformation. It is definitely known that variations of steel making practice reflect themselves in changes in the impact curves of nonembrittled specimens which will be reflected in changes in *apparent* temper brittleness as measured by ratios of room temperature impact test results. This variation of apparent temper brittleness has given rise to the widespread belief that temper brittleness is affected by steel making practice.

Manganese, chromium, and nickel, all definitely appear to increase the susceptibility, for without any of these elements (manganese less than about 0.60 per cent) steel is apparently not susceptible to temper brittleness. These elements probably increase the amount of transformation which can occur at each temperature (and indirectly the rate at which the precipitate forms). Manganese seemingly has a greater effect than chromium and chromium greater than nickel. Molybdenum decreases the amount of precipitate which occurs with moderate embrittling treatments. Although there may be evidence to the contrary, molybdenum appears to accomplish this effect by decreasing the rate of precipitation (but not the total amount which can occur). Molybdenum may be used therefore to ameliorate the embrittlement in moderately susceptible steels. In fact, with low manganese nickel-chromium steels 0.25 per cent molybdenum will prevent embrittlement even upon furnace cooling after tempering from above 600 degrees Cent. (1110 degrees Fahr.). There are some data which indicate that tungsten and possibly titanium and columbium affect the transformation in a manner similar to that of

molybdenum. It appears that the apparent effect of phosphorus in increasing the apparent temper brittleness is caused by an increase in the temperature of brittle failure of nonembrittled steels of high phosphorus content which in some cases brings the divergence in the impact curves between embrittled and nonembrittled specimens closer to room temperature. The effects of elements such as vanadium, nitrogen, and oxygen are not definitely known.

In the light of the interpretation of the effect of the temper brittleness transformation discussed in this paper, the older theories explaining its nature must be re-evaluated. Explanations based on phase transformation other than a simple precipitation have been suggested but the preponderance of data indicates that such hypotheses are not tenable.

A simple precipitation has been accepted by most students of the phenomenon as the transformation which results in the embrittlement. It has been variously suggested that the precipitate is a carbide, a phosphide, a nitride, or an oxide. The rate of diffusion of the responsible element is such, however, that either carbon or nitrogen appear to be involved; the facts point toward the latter element. It appears possible by deductive arguments to eliminate the precipitation of carbide as a possible cause of temper brittleness. A possible mechanism of the effects of alloying elements on the solubility of nitrogen in steel to account for the observed facts is suggested. Certainly any explanation at this stage must be tentative and justified by carefully designed experiments. It is suggested that experiments be performed to determine whether the precipitate is a (iron) carbide.

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Appendix A

Impact Tests on Susceptible Steels

1. Manganese-Vanadium Steel.

A sample of 5/8-inch rolled plate of a steel having the following composition:

Composition, Per Cent								
C	Mn	Si	S	P	Ni	Cr	Mo	Va
0.15	1.30	0.20	0.021	0.020	Nil.	0.025	.....	0.115

was machined into 5/8-inch square blanks. The blanks were austenitized at 900 degrees Cent. (1650 degrees Fahr.) for 1 hour and water-quenched. The structure after quenching consisted of martensite and bainite. The blanks were tempered at 655 degrees Cent. (1210 degrees Fahr.) for 1 hour, one set of specimens was furnace-cooled at 1/3 degree Cent. per minute after tempering and another set was water-quenched. Charpy specimens having a "V" notch of 0.01 inch radius were machined from the blanks. The results of hardness and impact tests are presented in Table AI. The hardnesses are the average of 6 readings and the impact energies are the result of only one measurement.

2. Manganese-Molybdenum Steel.

A sample of 3/4-inch thick rolled plate of a steel of the following composition:

Composition, Per Cent									
C	Mn	Si	S	P	Ni	Cr	Mo	Va	Ti
0.25	2.10	0.16	0.022	0.019	Nil.	0.04	0.53	0.05	0.05

was machined into 5/8-inch square blanks. The blanks were austen-

Table AI
Impact and Hardness Results from Manganese-Vanadium Steel

Embrittling Treatment	Average Hardness (Rockwell "B")	Specimen Number	Testing Temperature (Degrees Cent.)	Impact Energy (Ft.-Lbs.)
None	101	1	+200	125.5*
		2	+100	122.6*
		3	+ 20	131.7*
		4	− 40	127.0*
		5	− 60	20.8
		6	− 78	12.1
Furnace-Cooled	97	7	+200	131.3*
		8	+100	131.3*
		9	+ 20	127.9*
		10	− 10	109.7
		11	− 40	8.9
		12	− 78	5.6

*Specimens unbroken.

itized at 925 degrees Cent. (1695 degrees Fahr.) for 1½ hours and water-quenched. After quenching they were tempered at 665 degrees Cent. (1230 degrees Fahr.) for 1 hour, water-quenched and embrittled as indicated in Table AII. V-notched Charpy specimens were machined from the blanks and tested at the indicated temperatures. The results of hardness and impact tests are also presented in Table AII. The hardnesses are the average of 12 readings and the impact energies of 2 measurements.

3. Plain Carbon Steel.

A sample of ½-inch rolled plate of a steel of the following composition:

Composition, Per Cent									
C	Mn	Si	S	P	Ni	Cr	Mo	Va	Cu
0.395	0.85	0.29	0.018	0.009	Nil.	0.01	Trace	Nil.	0.085

was machined into ½-inch square blanks. The blanks were austen-

Table AII

Impact and Hardness Results from Manganese-Molybdenum Steel

Embrittling Treatment	Average Hardness (Rockwell "C")	Specimen Number	Testing Temperature (Degrees Cent.)	Average Impact Energy (Ft.-Lbs.)
None	29.5	6A	+200	91.9
		6B	+200	
		5A	+100	87.2
		5B	+100	
		4A	+ 20	80.4
		4B	+ 20	
		2A	− 40	77.2
		2B	− 40	
		1A	− 78	24.3
		1B	− 78	
½ Hr. at 475 Degrees Cent.	29.5	19A	+200	96.7
		19B	+200	
		18A	+100	87.5
		18B	+100	
		17A	+ 20	84.9
		17B	+ 20	
		16A	− 40	70.4
		16B	− 40	
		15A	− 78	20.5
		15B	− 78	
5 Hrs. at 475 Degrees Cent.	29.5	26A	+200	91.8
		26B	+200	
		25A	+100	81.1
		25B	+100	
		24A	+ 20	75.5
		24B	+ 20	
		23A	− 40	37.5
		23B	− 40	
		22A	− 78	20.3
		22B	− 78	
50 Hrs. at 475 Degrees Cent.	29.0	33A	+200	85.7
		33B	+200	
		32A	+100	72.3
		32B	+100	
		31A	+ 20	74.2
		31B	+ 20	
		30A	− 40	31.6
		30B	− 40	
		29A	− 78	16.5
		29B	− 78	

itized at 835 degrees Cent. (1535 degrees Fahr.) for 1 hour and air-cooled. They were then tempered at 655 degrees Cent. (1210 degrees Fahr.) for 1 hour. One set of the blanks was water-quenched and another furnace-cooled at ⅓ degree Cent., 1 minute. The hardnesses (average of 12 readings) and the impact energies (average of two values) are presented in Table AIII.

Table AIII
Impact and Hardness Results from Plain Carbon Steel

Embrittling Treatment	Average Hardness (Rockwell "B")	Specimen Number	Testing Temperature (Degrees Cent.)	Average Impact Energy (Ft.-Lbs.)
None	64	B-9	+200	42.2
		B-10	+200	
		B-7	+100	43.2
		B-8	+100	
		B-5	+ 20	43.0
		B-6	+ 20	
		B-11	0	24.0
		B-12	0	
		B-13	− 20	19.7
		B-14	− 20	
		B-3	− 40	10.5
		B-4	− 40	
		B-2	− 78	6.1
		B-1	− 78	
Furnace-Cooled	61	C-9	−200	47.5
		C-10	−200	
		C-7	−100	45.6
		C-8	−100	
		C-5	− 20	44.3
		C-6	− 20	
		C-11	0	26.9
		C-12	0	
		C-13	− 20	19.6
		C-14	− 20	
		C-3	− 40	12.6
		C-4	− 40	
		C-2	− 78	7.0
		C-1	− 78	

Appendix B

Impact Tests on Cast Steels

Specimens of 21 heats having the analyses listed in Table BI were heat treated in the form of ⅝-inch rounds. All of the specimens consisted of martensite after quenching and were cooled below room temperature to eliminate any retained austenite. The blanks of the various heats were tempered at temperature chosen to give Rockwell C-26 hardness. One set of the specimens from each heat was furnace-cooled at a rate of about ⅓ degree Cent. per minute and another set was water-quenched. After heat treatment V-

notched Charpy specimens were machined from the rounds and tested at —40 degrees Cent. The impact energies and hardnesses are listed in Table BII. The impact energies are the average of two specimens and the hardnesses are the average of 12 readings. The results are plotted in Figs. 32 and 33 of the text.

Table BI
Chemical Analyses and Heat Treatments of Cast Steel

Heat No.	Composition Per Cent C	Mn	Si	S	P	Ni	Cr	Mo	Res. Al	Aust. Temp. (4 Hrs.) Deg. Fahr.	Tempering Temp. (4 Hrs.) Deg. Fahr.
1	0.315	1.01	0.22	0.016	0.014	0.32	0.48	0.35	0.02	1650	1155
2	0.325	1.01	0.26	0.021	0.013	0.32	0.79	0.37	0.03	1650	1175
3	0.31	1.12	0.33	0.018	0.011	0.37	1.18	0.33	0.03	1650	1205
4	0.35	1.31	0.23	0.017	0.014	0.32	0.45	0.37	0.02	1650	1165
5	0.27	1.22	0.24	0.020	0.014	0.34	0.74	0.37	0.025	1650	1155
6	0.33	1.32	0.40	0.023	0.012	0.34	1.18	0.33	0.03	1650	1210
7	0.32	1.40	0.38	0.019	0.014	0.35	1.45	0.375	0.025	1700	1220
8	0.32	1.52	0.42	0.020	0.014	0.41	1.22	0.37	0.025	1700	1220
9	0.32	1.52	0.38	0.017	0.013	0.36	1.44	0.39	0.025	1725	1210
10	0.32	1.32	0.44	0.023	0.012	1.10	1.68	0.49	0.04	1675	1225
11	0.265	1.37	0.37	0.017	0.010	1.12	2.17	0.49	0.04	1725	1180
12	0.285	1.33	0.32	0.016	0.012	2.27	1.63	0.44	0.025	1725	1200
13	0.255	1.31	0.37	0.016	0.012	2.03	2.10	0.51	0.035	1725	1195
14	0.279	1.32	0.36	0.016	0.012	3.07	1.56	0.52	0.035	1725	1230
15	0.29	1.35	0.38	0.016	0.012	3.29	2.14	0.49	0.035	1725	1230
16	0.27	0.76	0.32	0.017	0.010	3.29	1.54	0.49	0.03	1725	1205
17	0.34	0.92	0.37	0.019	0.009	3.13	1.68	0.43	0.05	1650	1225
18	0.33	1.12	0.36	0.019	0.011	3.41	1.71	0.46	0.025	1650	1205
19	0.31	0.76	0.35	0.018	0.010	tr.	2.57	0.45	0.035	1700	1210
20	0.39	1.33	0.39	0.016	0.013	3.31	1.58	0.49	0.035	1725	1205
21	0.375	1.39	0.39	0.017	0.013	2.99	2.16	0.48	0.025	1725	1205

Table BII
Impact (At —40 Degrees Cent.) and Hardness Results for Cast Steels

Heat No.	Furnace-Cooled After Tempering Impact Energy (Ft.-Lbs.)	Hardness Rockwell "C"	Oil-Quenched After Tempering Impact Energy (Ft.-Lbs.)	Hardness Rockwell "C"
1	34.3	24.2	31.9	26.2
2	44.6	24.9	49.9	25.5
3	44.6	25.0	49.2	25.2
4	35.0	24.9	35.2	25.6
5	37.5	24.5	38.0	26.5
6	31.0	26.8	43.3	26.9
7	28.8	26.8	47.0	27.5
8	17.3	26.1	43.5	27.4
9	26.5	26.1	48.8	26.6
10	24.3	25.1	46.7	26.8
11	6.8	25.7	38.7	25.7
12	8.6	25.7	40.4	25.5
13	5.3	25.3	32.0	25.4
14	10.6	24.9	45.9	25.0
15	5.0	23.6	51.1	24.3
16	17.9	25.3	57.1	24.6
17	26.0	24.7	48.1	26.7
18	14.2	29.0	43.8	27.5
19	4.5 0	26.8	54.7	28.5
20	6.6	30.9	23.1	28.4
21	3.4	28.8	17.7	28.6

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DISCUSSION

HAROLD OSTER:[24] I believe, if I recall correctly, the tendency for temper brittleness depended upon certain elements and the author said that nitrogen and manganese are eliminated. I presume that other researches are going on. May I ask him to give a guess as to what he believes is the cause of temper brittleness and to discuss what other researches are being done to eliminate other elements such as these.

GEO. H. ENZIAN:[25] I am interested in the author's conclusion that nitrogen below 0.001 per cent was not a factor in temper brittleness. I would like to ask Captain Hollomon how he made his steel with less than 0.001 per cent nitrogen and how he determined that that was the content present; that is, whether that was the total nitrogen or the soluble nitrogen.

H. H. CHISWIK:[26] In view of the fact that alloying elements are usually necessary to cause and increase the effects of temper brittleness, I wonder whether the author would care to discuss the possibilities of retained austenite as being one of the causes in some cases.

H. W. HIGHRITER:[27] It seems advisable to bring to the attention of this discussion the fact that the manifestations of temper brittleness are not limited

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[27]Chief metallurgical engineer, Vascoloy Ramet Corp., North Chicago, Ill.

necessarily to steels. A good many years ago, it was a very serious problem in the manufacture of malleable castings that were to be galvanized or tinned, and in some work done a number of years ago by E. S. Davenport and myself it was found that the susceptibility of malleable castings to embrittlement was quite definitely a function of the total of the silicon and phosphorus contents. In that case the susceptibility to embrittlement could be reduced by the addition of small amounts of molybdenum. It could also be eliminated by preliminary heat treatment, quenching of the castings from a temperature of about 1200 degrees Fahr. (650 degrees Cent.). I suggest that consideration be given to the possibility of some relationship between silicon and phosphorus contents of steels and their susceptibility to temper brittleness.

A. E. Focke:[28] Early in the War, I recall reading about the experiences of a man in England[29] who was heat treating parts similar to wheels. He was required to quench these parts after tempering because in one case the British Ordnance specifications required that the parts be so treated while in the other an impact test was specified and it was necessary to quench in order to meet the specified values.

But it was his experience that parts so treated were so highly stressed as the result of the quench that they broke in service much earlier than did similar parts which he had purposely cooled slowly after the tempering. I would appreciate Captain Hollomon's comment on this observation.

Lloyd R. Jackson:[30] I would like to ask if the longitudinal fracture that was observed as one indication of temper brittleness is observed on cast test pieces as well as on forged or worked ones?

Author's Closure

The author would like to express his appreciation of the comments of those who discussed his paper. The questions by Mr. Oster and Mr. Enzian can be answered simultaneously. To date, several heats of steel have been made by vacuum melting under pressures of the order of 10^{-4} mm. of mercury. When analyzed for nitrogen by chemical methods (for total nitrogen) they were found to contain less than 0.002 per cent. The accuracy of these chemical methods is, as far as the author is concerned, anyone's guess. The steels contain, when processed in this way, less than 0.01 per cent manganese. Additional nickel was added to counteract the loss of manganese. If these experiments demonstrate anything, they demonstrate that manganese itself is not responsible for temper brittleness, although it is well known that it increases the susceptibility. It is further certain that if nitrogen is responsible, only minute amounts cause the difficulty.

No further researches are at this moment planned to determine the cause of temper brittleness. It is now believed very likely that carbon itself may be responsible. If this is found to be the case, then of course steels as we know them will be made free of temper brittleness only with great difficulty.

In answer to Mr. Chiswik, the author would like to call attention to the paper by Esser and Eilander (55) who suggested the mechanism of decomposition of

[28]Chief metallurgist, Diamond Chain & Mfg. Co., Indianapolis.

[29]L. E. Benson, "The Quenching of Steel After Tempering and the Impact Test," *Engineering*, August 1942, p. 134.

[30]Research supervisor, Battelle Memorial Institute, Columbus, Ohio.

austenite as the cause for temper brittleness. If austenite decomposition were the cause, then the brittleness could not be made to appear and reappear in the same steel without exceeding the lower critical temperature. This behavior is a characteristic of the temper brittleness phenomenon. It is also possible to reduce the impact properties of steels by means of decomposition of retained austenite. However, once the austenite is decomposed, quenching from the subsequent tempering will not renew the old properties.

With regard to the effect of silicon and phosphorus, Mr. Highriter is referred to the discussion concerning phosphorus (page 508). It is extremely unlikely that either silicon or phosphorus is directly responsible for temper brittleness, although they may contribute to enhancing its apparent effect by lowering the properties of the unembrittled steel.

Certainly the manufacturer faces a dilemma as pointed out by Mr. Focke when attempting to eliminate the effects of temper brittleness in a susceptible steel. It is possible that unfavorable residual stresses may be introduced upon quenching following tempering, although in simple shapes such an unfavorable stress distribution is unlikely. If the steel is not quenched following tempering, temper brittleness is liable to result. However, residual stresses in themselves do not induce service failures in steels of satisfactory toughness. Other causes must have been associated with the failure reported by Benson.

In answer to Mr. Jackson, the author has not observed longitudinal fractures in cast steels. However, it would be expected that specimens taken parallel to the forging direct would be more susceptible to this type of fracture than would specimens taken from castings. That the forging is not the primary cause of this fracture can easily be demonstrated. Two specimens taken from the same forging can be pulled in simple tension. One of these embrittled by the temper brittleness precipitation will exhibit the longitudinal fracture. The other specimen having identical physical properties in simple tension will break with the normal cup cone fracture.

HIGH FORGING TEMPERATURES REVEALED BY FACETS IN FRACTURE TESTS

By J. Robert Strohm and W. E. Jominy

Abstract

The appearance of large grains or facets in the fracture of alloy steel forgings is an indication of questionable quality and of high temperature forging. Temperatures below which facets are not formed were accurately determined by means of thermocouples inserted in test pieces and by fracture testing production forgings. Facets appear in fractures of quenched and tempered forgings and not in normalized or annealed forgings. They are difficult to find unless the fracture is perpendicular to the direction of grain flow. The tensile properties of bars containing facets are found to be less favorable than those which have no facets.

THE term "facets" is rapidly becoming common around forge shops, since fracturing of test lugs has become a production operation on certain forgings. This fracturing of test lugs is applied only to those few forgings which must give the utmost in performance. It has usually been made part of the test procedure as a result of service failures in which high forging temperatures appear to have been a contributing factor in the failure. Facets appearing in the fracture are regarded as an indication of questionable quality.

Facets are usually associated with overheating for forging. They occur in the fracture of the overheated section, being light in color and stand out clearly from the background of darker shades. They vary in orientation from one grain to another, and although the facet is angular, its surfaces look smooth. It has the appearance of crystalline material that has been broken along definite cleavage planes. The photograph, Fig. 4, of S.A.E. 4640 steel heated to 2500 degrees Fahr. (1370 degrees Cent.) shows a fracture containing many facets.

Facets are seen in fractures of parts which have been quenched

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and tempered and do not appear if the steel is in the as-forged, normalized, or annealed condition. They appear only if the fracture is transverse to the grain flow and are visible without the use of a magnifying glass. When facets occur in the fracture of a forging, the part is thought to be not acceptable; thus, it becomes the problem of the drop forge plant to overcome the conditions that cause them.

To study the conditions under which facets are formed, we made a series of tests in which steel samples were heated to various forging temperatures, air-cooled, normalized, hardened, tempered, and fractured. Five steels were investigated and it was found that for each steel, except nitralloy G, a temperature existed below which no facets were produced, and above which facets were always produced. Such a temperature may also exist for nitralloy G if high enough temperatures are employed. In this work, the highest temperature used was 2500 degrees Fahr. (1370 degrees Cent.) and at this temperature, nitralloy G showed no facets. It was found that facets were not produced at the same temperature for all analyses of steel. Although samples for the study were held 15 minutes at temperature, several other samples were placed in a furnace and held 5 hours at temperature, and it was found that this extra time at temperature made no appreciable difference in the size or number of facets. Forging at temperatures below the facet forming temperature eliminated the facets, if enough work was done.

The steels used for the study were: Nitralloy G modified, S.A.E. steels 3312, 4140, 4340 and 4640. Specimens for test were cut approximately 2 inches square and 12 inches long, and were taken from a quarter section of the bar or billet. In heating the samples, extreme care was taken to be certain of the temperature used. All samples were drilled with a ⅜-inch hole, 5 inches deep, and counterbored with a 1⅜-inch hole, 2 inches deep, so that the end of a black iron pipe, 4 feet long, could be inserted in the end of the sample. A chromel-alumel couple was then inserted through the pipe into the sample and the other end was connected to a potentiometer for temperature readings. Each couple was checked against a standard couple before and after each test and was found to be satisfactory. Table I gives analysis and heat treatment for each steel studied.

Although all quenched structures may be used to reveal facets, we chose a hardness of about 300 Brinell, as it is easier to machine tensile bars and Charpy specimens at this hardness; also, when high tempering temperatures are used to make the specimens lower than

Table I
Composition and Heat Treatment Used for Each Grade of Steel Studied

Steel	Analysis									Heat Treatment, Degrees Fahr.			
	C	Mn	P	S	Si	Cr	Mo	Ni	Al	Normalize Air-Cooled Below 600° F.	Harden Oil Quench	Temper	Brinell No.
SAE 4140 AMS 6382	0.40	1.01	0.01	0.02	0.28	1.07	0.23	....	....	1725	1550	1150 3 Hrs.	302
SAE 4640 AMS 6312	0.42	0.70	0.015	0.017	0.27	....	0.23	1.88	....	1650	1525	1025 3 Hrs.	311
SAE 4340 AMS 6412	0.38	0.75	0.015	0.017	0.26	0.83	0.23	1.83	....	1650	1525	1150 3 Hrs.	311
SAE 3312 AMS 6250	0.11	0.59	0.02	0.009	0.30	1.67	....	3.32	....	1725	1600	600 3 Hrs.	311
Nitralloy G Modified	0.42	0.63	0.01	0.016	0.28	1.62	0.33	....	1.07	1800	1700	1175 5 Hrs.	311

300 Brinell, it becomes difficult to get a clean fracture and a shearing action may obscure the test.

The samples for the study were placed in the furnace at different temperatures with their respective couples as previously described, and held 15 minutes at heat after the couples showed the samples to be at temperature. Half the samples were then air-cooled, normalized, quenched, and tempered; the other half were forged from 2-inch squares into 1-inch rounds on a blacksmith hammer and finish forged at 1800 degrees Fahr. (980 degrees Cent.). These were then air-cooled, normalized, quenched, and tempered. All samples were fractured and photographs were made of them.

In Fig. 1 photographs of the fractures of the samples of S.A.E. 4140 steel are shown. The samples that were heated to 2350 and 2400 degrees Fahr. (1290 and 1315 degrees Cent.) show no evidence of facets; the sample that was heated to 2450 degrees Fahr. (1345 degrees Cent.) shows facets rather faintly while the sample heated to 2500 degrees Fahr. (1370 degrees Cent.) shows facets more prominently. The fractures shown in all the photographs were obtained after heat treatment as given in the previous table. The photographs, Fig. 2, of the fractures of S.A.E. 3312 of samples that were heated to 2350 and 2400 degrees Fahr. (1290 and 1315 degrees Cent.) show no facets; but the sample that was heated to 2480 degrees Fahr. (1360 degrees Cent.) shows them quite clearly. The photograph of the fracture of S.A.E. 4340 that was heated to 2350 degrees Fahr. (1290 degrees Cent.) shows no facets; however, the other three temperatures, 2400, 2450, and 2500 degrees Fahr. (1315, 1345 and 1370 degrees Cent.), all produced facets most distinctly, as shown in Fig. 3. The photograph of the fracture of S.A.E. 4640 that was heated to 2350 degrees Fahr. (1290 degrees Cent.) shows no facets; however, the succeeding temperatures of 2400 to 2500 degrees Fahr. (1315 to 1370 degrees Cent.) show facets which become more distinct as the temperature is increased, as presented in Fig. 4. The photograph of the fracture of nitralloy G, presented in Fig. 5, shows no facets occurring at any temperature.

The photographs, Fig. 6, of the fractured samples heated to high temperatures and forged show no facets. Upon reheating these samples to 2500 degrees Fahr. (1370 degrees Cent.) and air cooling without forging and followed by their usual heat treatment, facets appeared. These samples were then again heated to 2500 degrees Fahr. (1370 degrees Cent.) and forged, after which they were nor-

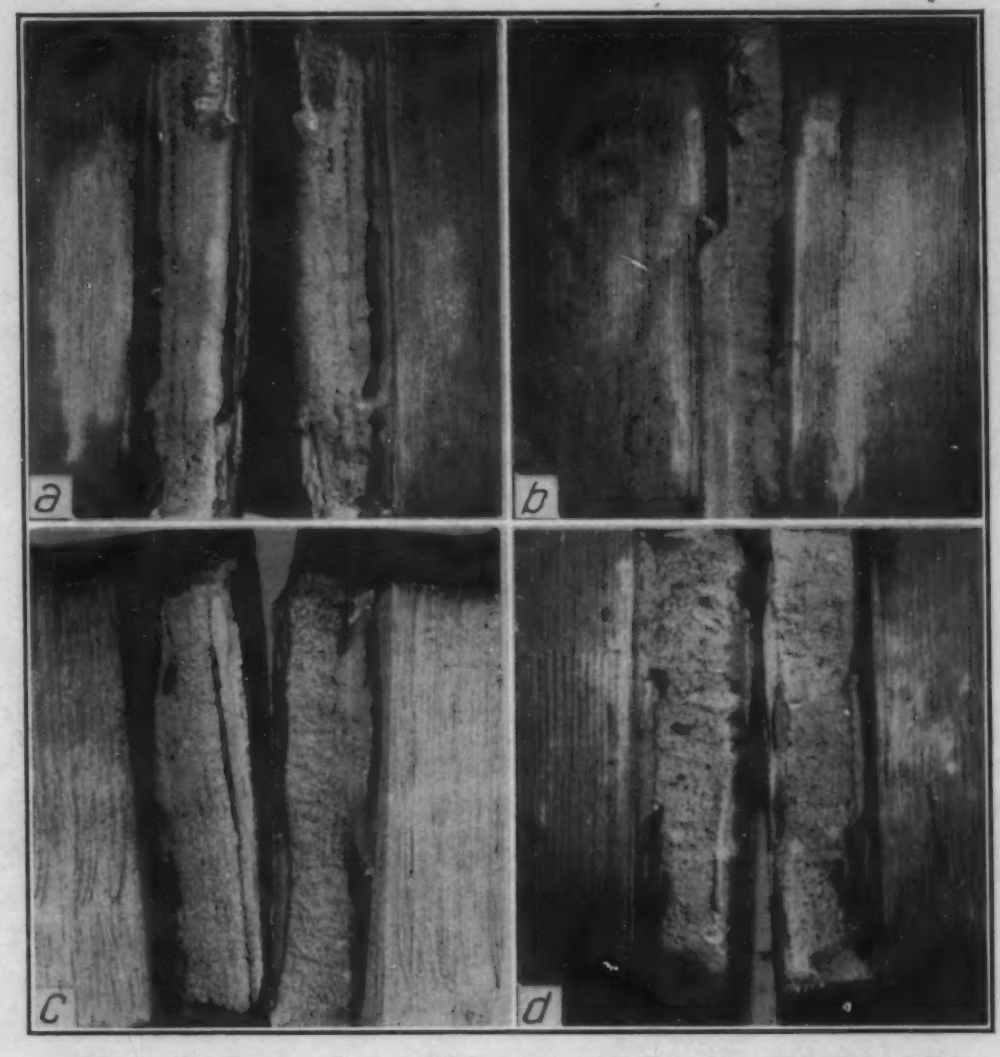

Fig. 1—S.A.E. 4140 Steel (A.M.S. 6382). (a) Heated to 2350 degrees Fahr. (b) Heated to 2400 degrees Fahr. (c) Heated to 2450 degrees Fahr. Facets beginning. (d) Heated to 2500 degrees Fahr. Facets. All specimens, × 2.

malized, quenched, tempered, and fractured, and no facets were found in the fractures. It thus appears that facets can be eliminated by forging provided the forging is done at temperatures below the facet forming temperatures.

All samples for this study were heated in an oil-fired conventional door type forge furnace. Although the thermocouple checks in this furnace were accurate, we wished to further confirm our results in the laboratory. To do this, we used a globar electric furnace whose temperature was controlled by means of a platinum thermocouple connected to a potentiometer recorder. The samples for the laboratory were prepared the same as for the shop tests. The base

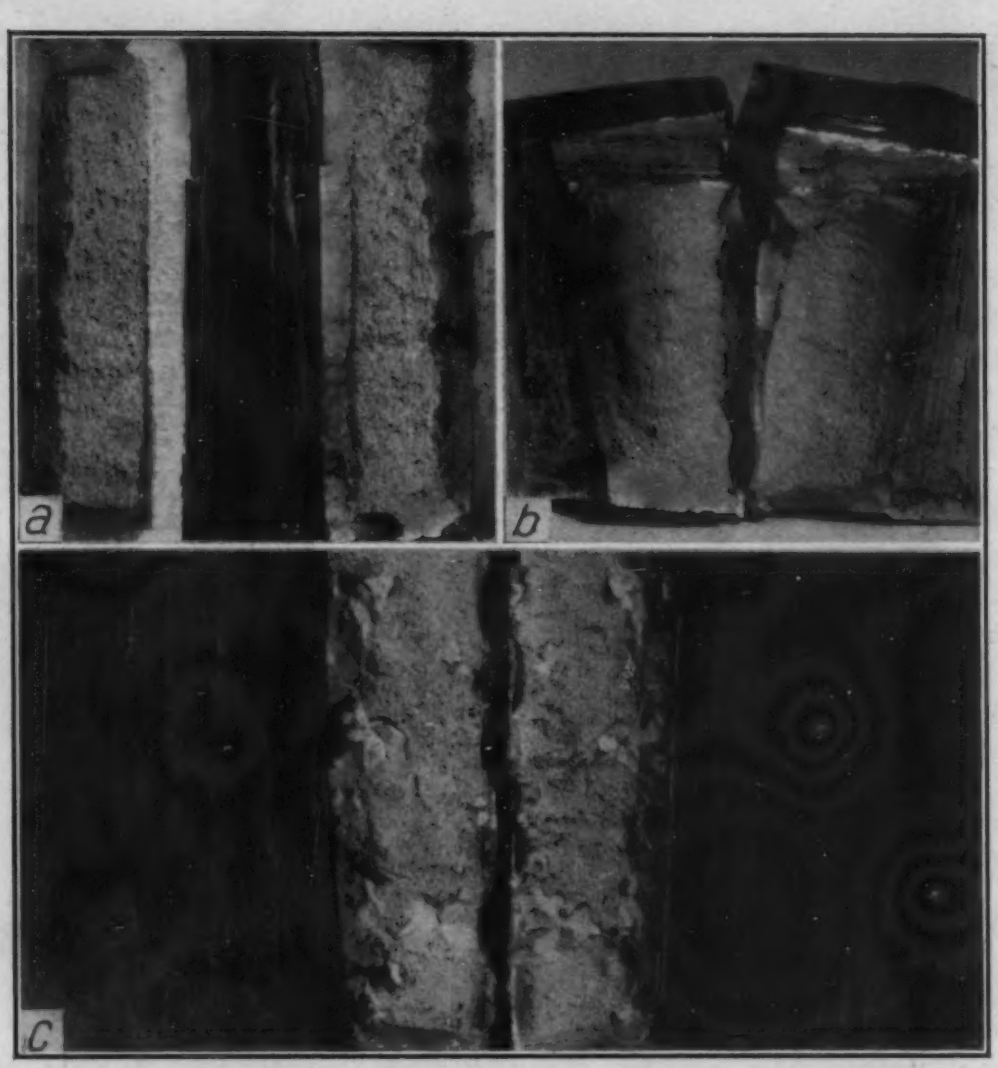

Fig. 2—S.A.E. 3312 Steel (A.M.S. 6250). (a) Heated to 2350 degrees Fahr. (b) Heated to 2400 degrees Fahr. (c) Heated to 2480 degrees Fahr. Facets. All specimens, × 2.

metal thermocouples that were inserted in the laboratory pieces were checked before and after the test against an optical pyrometer while submerged in the target tube of a radiation pyrometer installed in a high speed steel heating furnace. The optical pyrometer had been previously calibrated against a standard lamp. With this equipment, of course, we were able to obtain closer control. We heated S.A.E. 4340 and S.A.E. 4640 steel to 2375 degrees Fahr. (1300 degrees Cent.) and then heat treated as previously described and no facets appeared, but upon heating these two steels to 2400 degrees Fahr. (1315 degrees Cent.) followed by their usual heat treatment, facets appeared. We further found that no facets ap-

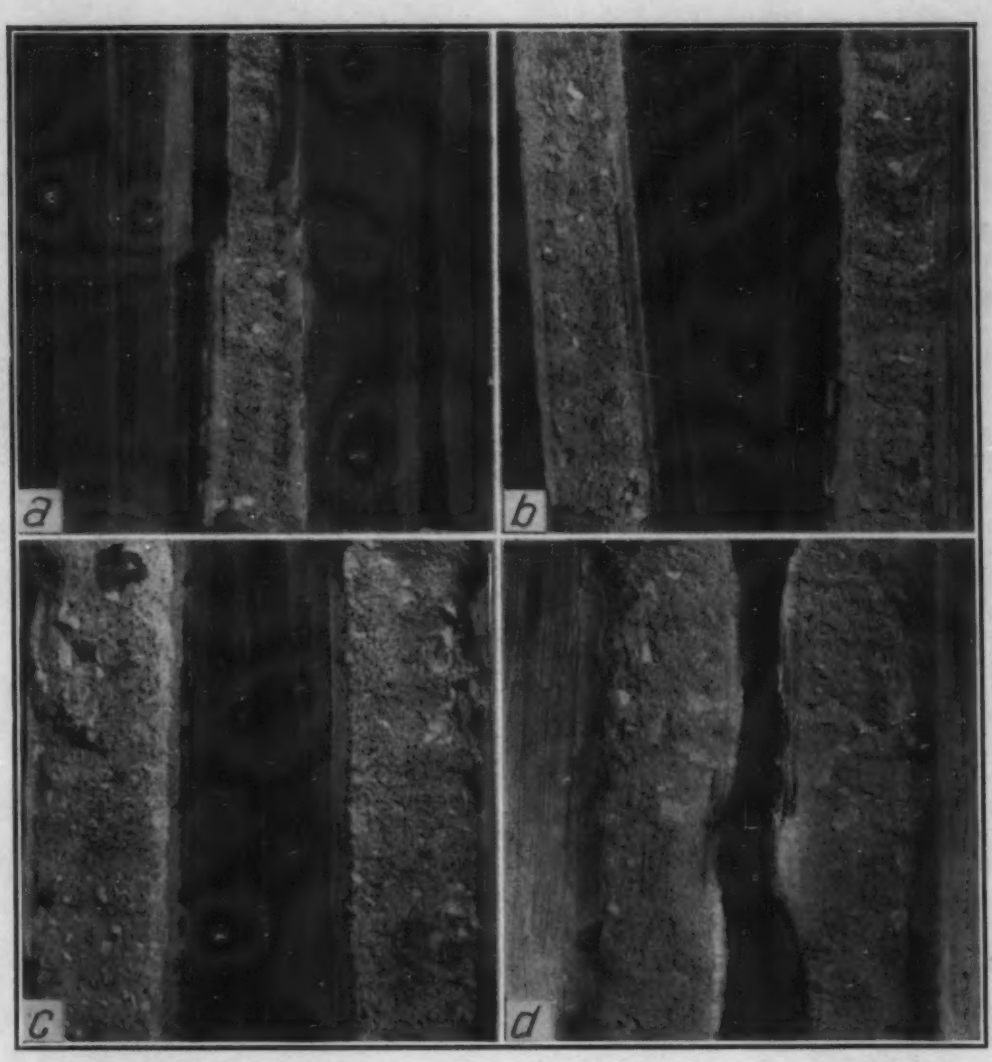

Fig. 3—S.A.E. 4340 Steel (A.M.S. 6412). (a) Heated to 2350 degrees Fahr. (b) Heated to 2400 degrees Fahr. Facets. (c) Heated to 2450 degrees Fahr. Facets. (d) Heated to 2500 degrees Fahr. Facets. All specimens, × 2.

peared upon heating S.A.E. 3312 and S.A.E. 4140 steel to 2425 degrees Fahr. (1330 degrees Cent.) followed by their usual heat treatment, but that upon heating to 2450 degrees Fahr. (1345 degrees Cent.) and heat treating, facets appeared. We were, therefore, able to determine the highest temperature at which these steels may be heated without the formation of facets.

In addition to these tests, fractures have been made of production forgings in connection with quality control work. In excess of fifteen hundred fractures representing more than one hundred twenty-five heats of steel have been examined. As a result of this work, the following summation has been prepared, showing the highest

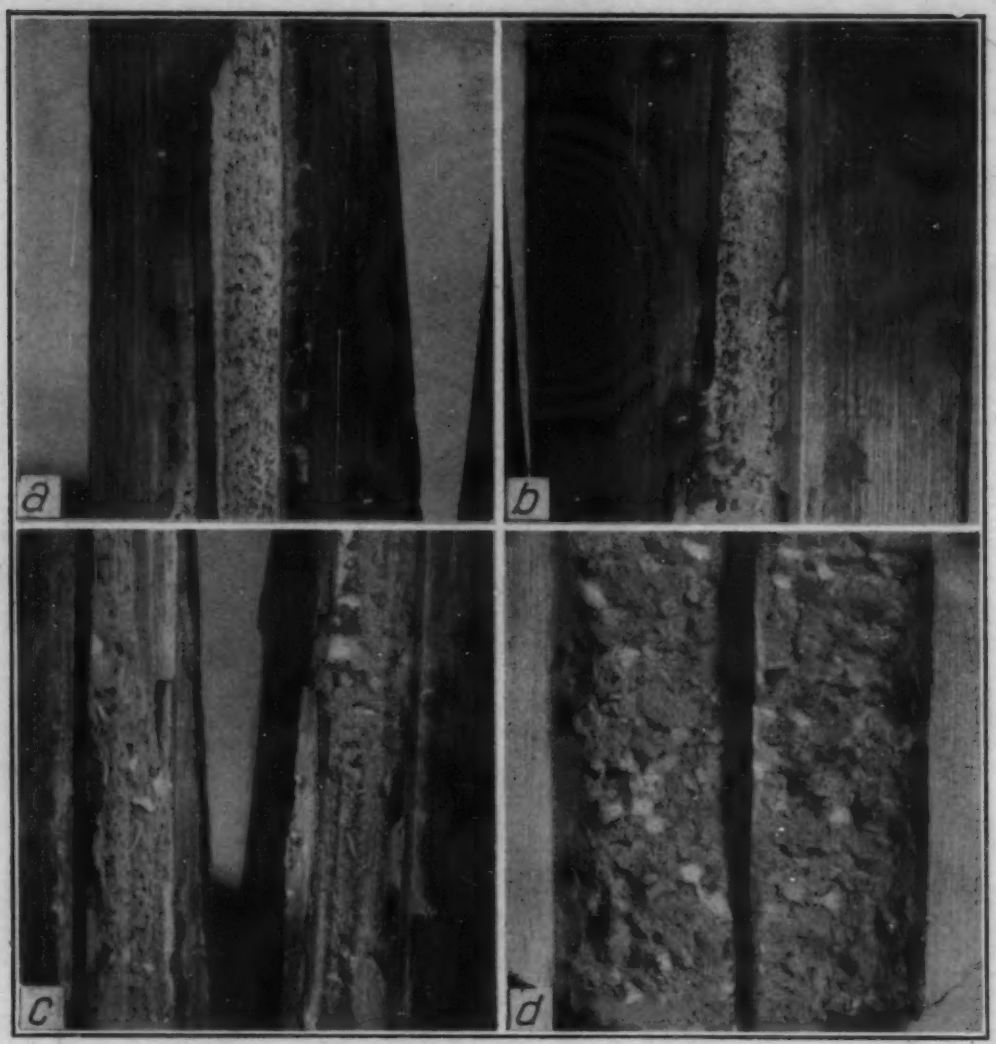

Fig. 4—S.A.E. 4640 Steel (A.M.S. 6312). (a) Heated to 2350 degrees Fahr. (b) Heated to 2400 degrees Fahr. Facets. (c) Heated to 2450 degrees Fahr. Facets. (d) Heated to 2500 degrees Fahr. Facets. All specimens, × 2.

temperature to which the steel may be heated without the formation of facets:

	Degrees Fahr.
S.A.E. 4340	2375
S.A.E. 4640	2375
S.A.E. 3312	2425
S.A.E. 4140	2425
Nitralloy G	2500

In the production of a certain forging, we found it possible to have the correct heating temperature and still form facets. This was due to the heat gained during the fullering and blocking operations. As measured by the optical pyrometer, the part actually gained 50 to 60 degrees Fahr. It is, consequently, important to keep the tempera-

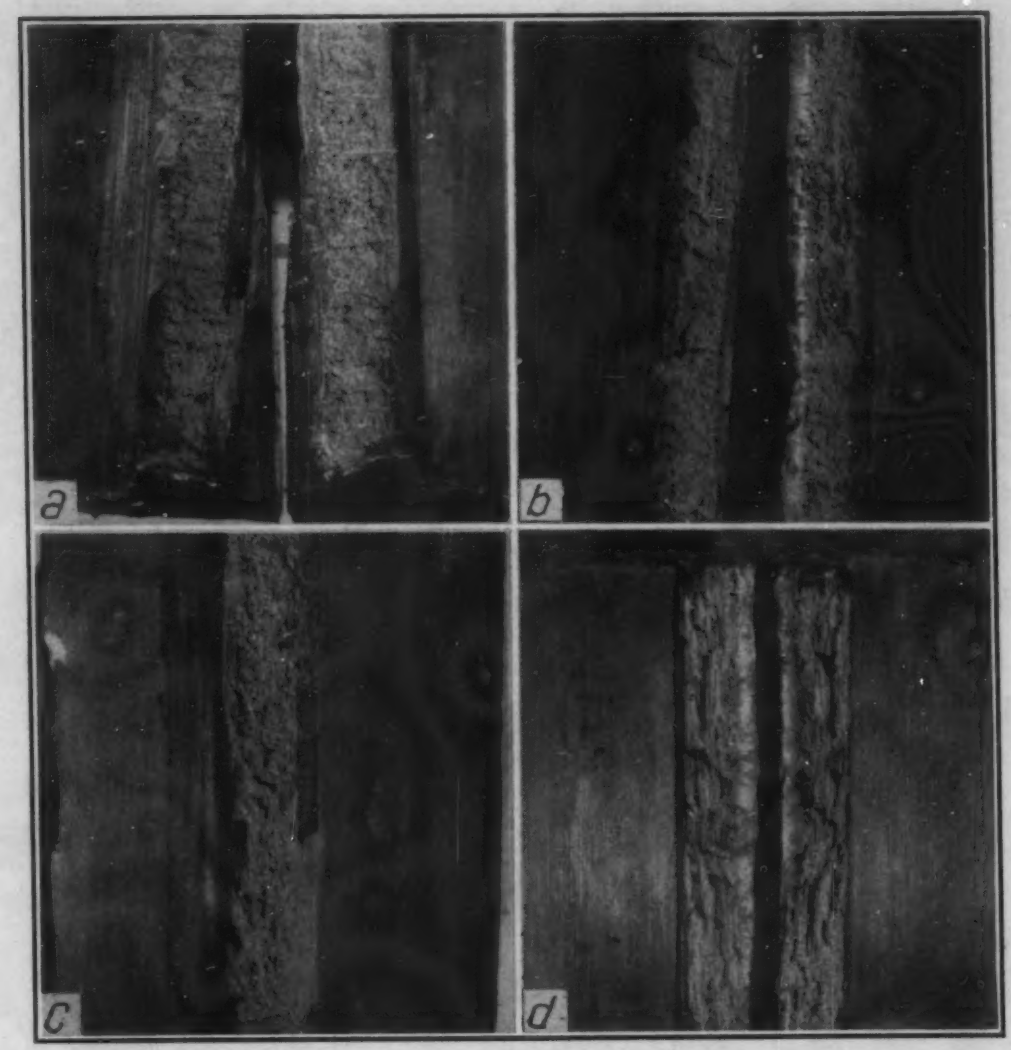

Fig. 5—A.M.S. 6470 Steel (Nitralloy G Modified). (a) Heated to 2350 degrees Fahr. (b) Heated to 2400 degrees Fahr. (c) Heated to 2450 degrees Fahr. (d) Heated to 2480 degrees Fahr. All specimens, × 2.

ture at which the bars are heated for forging well below the temperatures shown above.

After learning how facets can be produced or prevented, it was decided to determine what effect these facets had on the physical properties of the steel in which they occur. Tensile bars and Charpy specimens were made from the bars that were heated to temperatures which formed facets and others heated to just below the facet forming temperatures. These bars were of the compositions listed in Tables I through V, and were heat treated as indicated in these tables after the high temperature treatment. The physical properties obtained from these bars are presented in Tables II through V. As

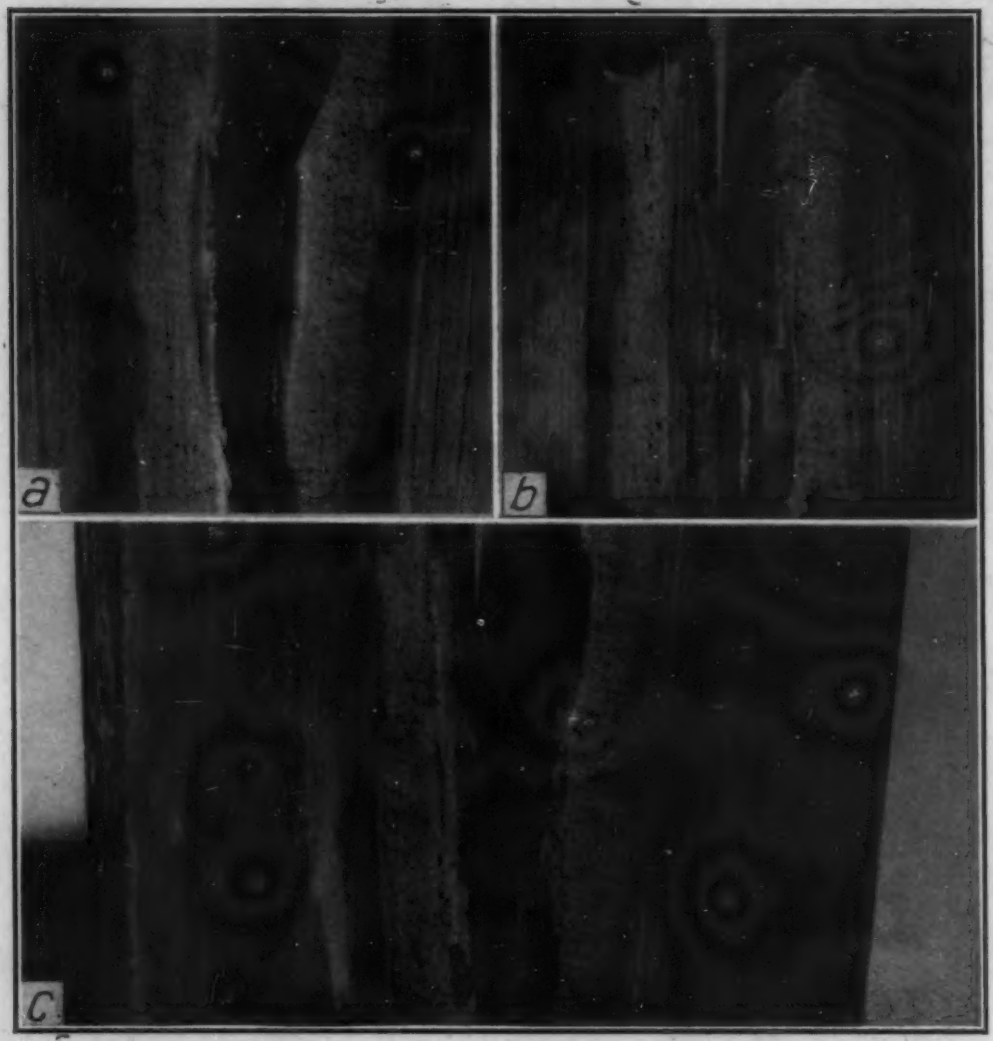

Fig. 6—(a) S.A.E. 3312 Steel (A.M.S. 6250) Heated to 2480 Degrees Fahr. Forged. (b) S.A.E. 4640 steel (A.M.S. 6312) heated to 2500 degrees Fahr. Forged. (c) S.A.E. 4340 steel (A.M.S. 6412) heated to 2500 degrees Fahr. Forged. All specimens, × 2.

may be observed, the impact tests revealed little or no conclusive evidence that the physical properties were impaired in any of the steels. The elongation and reduction of area results on each steel that had facets were lower than when no facets were present, indicating that higher temperatures have a deleterious effect on these properties. We made tensile bars only for temperatures of 2350 and 2500 degrees Fahr. (1290 and 1370 degrees Cent.) in this study. There was such a difference in results on elongation and reduction of area that additional tensile bars were made from S.A.E. 4640 that had been heated to 2400 degrees Fahr. (1315 degrees Cent.) to determine whether

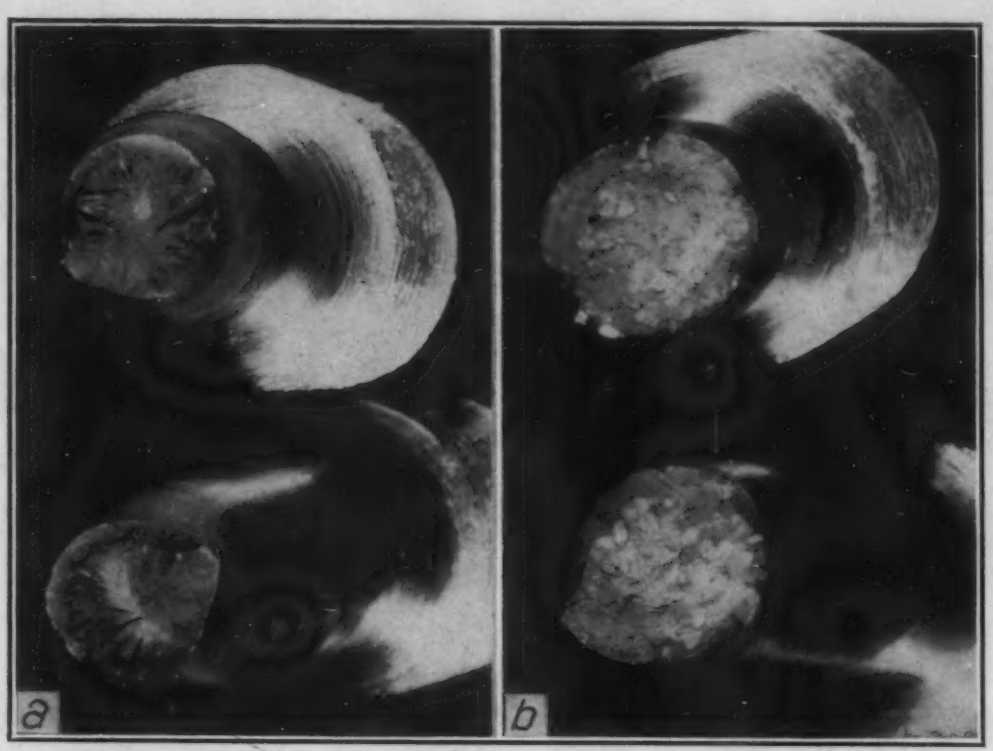

Fig. 7—Tensile Fractures of S.A.E. 4640 Steel (A.M.S. 6312). (a) Heated to 2350 degrees Fahr. (b) Heated to 2500 degrees Fahr. Both specimens, × 2.

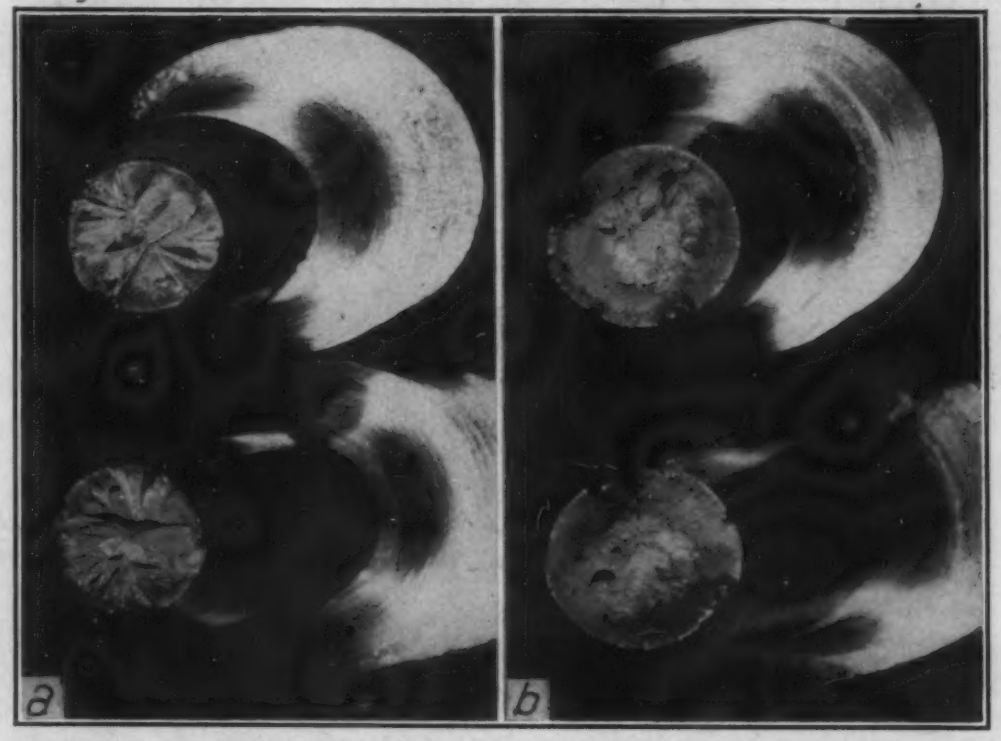

Fig. 8—Tensile Fractures of S.A.E. 4140 Steel (A.M.S. 6382). (a) Heated to 2350 degrees Fahr. (b) Heated to 2500 degrees Fahr. Both specimens, × 2.

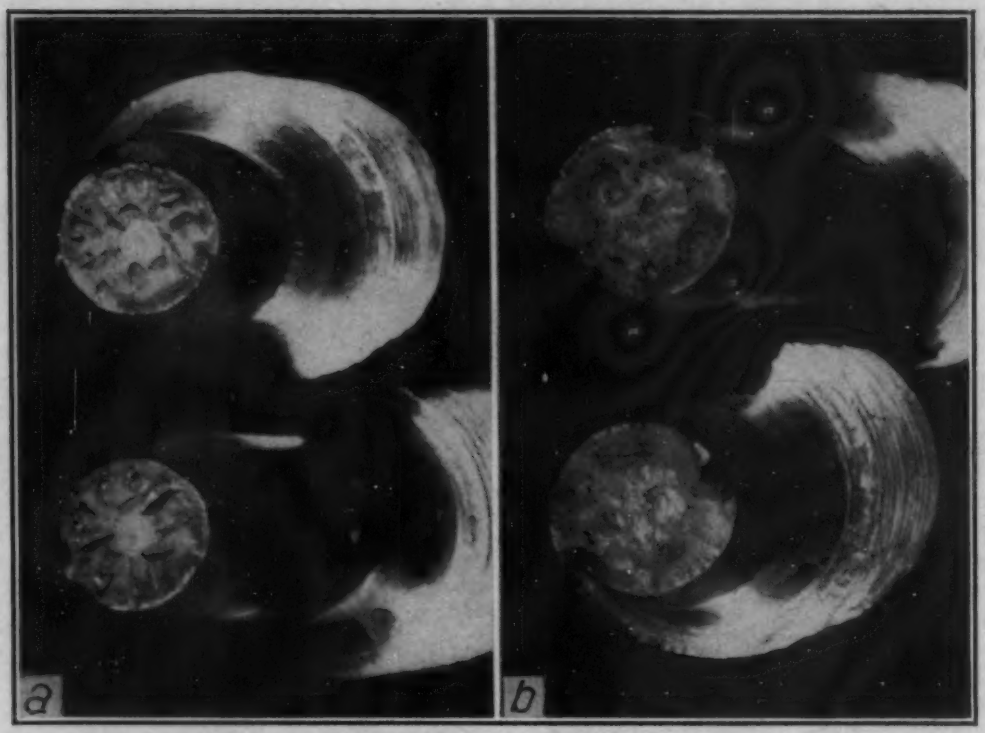

Fig. 9—Tensile Fractures of S.A.E. 4340 Steel (A.M.S. 6412). (a) Heated to 2350 degrees Fahr. (b) Heated to 2500 degrees Fahr. Both specimens, × 2.

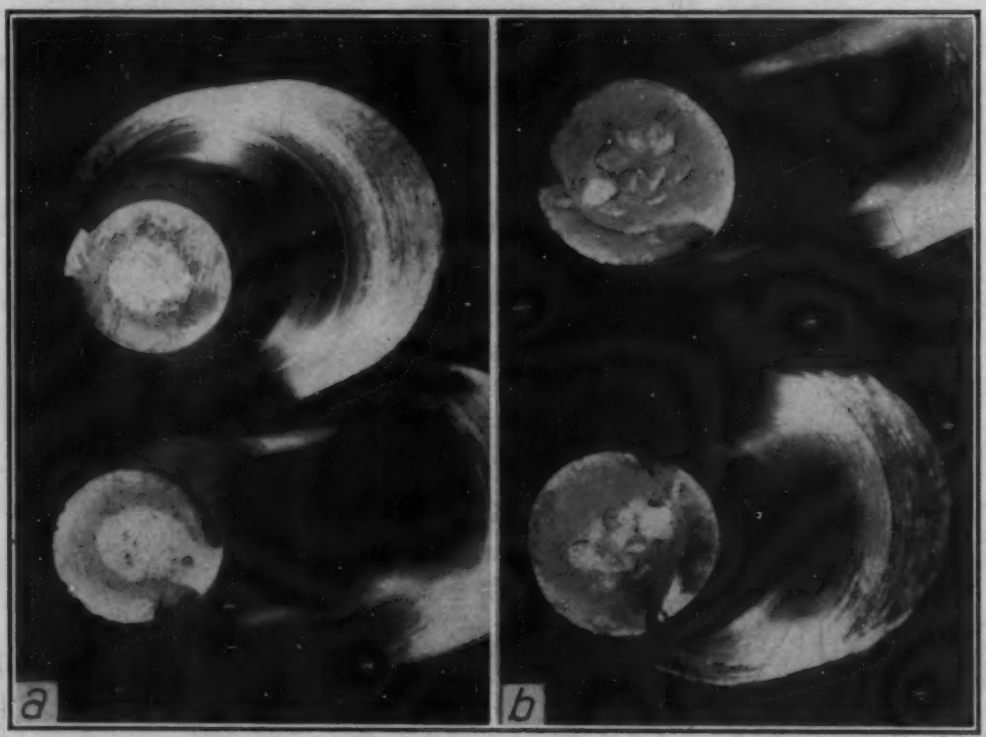

Fig. 10—Tensile Fractures of S.A.E. 3312 Steel (A.M.S. 6250). (a) Heated to 2350 degrees Fahr. (b) Heated to 2500 degrees Fahr. Both specimens, × 2.

Table II

Material SAE	Material AMS	Heating Temp. Degrees Fahr.	T. S. p.s.i.	Yield Point p.s.i.	Elong. Per Cent in 2 Inches	Red. Area Per Cent	Brinell	Charpy Impact Ft.-Lbs.
4640	6312	2350	153,500	136,000	19	58.5	321	49
4640	6312	2350	153,500	137,300	18	56.5	311	49
4640	6312	2350	151,800	138,500	19	58	311	49
Average			152,933	137,266	18.6	57.6	314	49
4640	6312	2400	146,200	125,000	17	55.4	311	..
4640	6312	2400	146,000	127,500	16	55.5	311	..
4640	6312	2400	145,000	123,000	17	56.2	311	..
4640	6312	2400	148,000	130,000	19	54.1	311	..
Average			146,300	126,375	17.2	55.3	311	..
4640	6312	2500	153,800	137,200	15	38	321	50
4640	6312	2500	156,500	145,000	17	53	321	51
4640	6312	2500	158,000	146,000	13	32.5	311	53
4640	6312	2500	157,000	144,000	16.5	50	311	50
Average			156,325	143,050	15.3	43.3	316	51

Table III

Material SAE	Material AMS	Heating Temp. Degrees Fahr.	T. S. p.s.i.	Yield Point p.s.i.	Elong. Per Cent in 2 Inches	Red. Area Per Cent	Brinell	Charpy Impact Ft.-Lbs.
4140	6382	2350	141,500	130,000	20	62	302	37
4140	6382	2350	143,500	123,000	21	60	302	37
4140	6382	2350	142,000	130,000	20	60	302	38
4140	6382	2350	141,300	129,000	21	59	302	37
Average			142,075	128,050	20.5	60.2	302	37
4140	6382	2500	144,500	133,500	20	58	302	38
4140	6382	2500	148,700	137,500	20	52	302	36
4140	6382	2500	147,500	136,500	19	54	302	37
4140	6382	2500	142,500	131,500	18	47	302	37
Average			145,800	134,750	19.2	52.7	302	37

Table IV

Material SAE	Material AMS	Heating Temp. Degrees Fahr.	T. S. p.s.i.	Yield Point p.s.i.	Elong. Per Cent in 2 Inches	Red. Area Per Cent	Brinell	Charpy Impact Ft.-Lbs.
4340	6412	2350	137,200	124,200	20	62	302	46
4340	6412	2350	136,300	122,500	20	60	302	41
4340	6412	2350	139,400	127,000	22	62	302	43
4340	6412	2350	140,000	127,800	21	62	302	44
Average			138,225	125,375	20.7	61.5	302	43.5
4340	6412	2500	136,900	124,300	17	53	302	38
4340	6412	2500	141,000	128,800	20	57	302	39
4340	6412	2500	140,000	127,400	20	58.5	302	38
4340	6412	2500	140,000	127,000	20	57	302	37
Average			139,475	126,875	19.2	56.3	302	38

the results on elongation and reduction of area fell in the middle of the high and low temperature results. This was found to be the case as indicated in Table II, which shows that as the temperature was increased, the reduction of area and elongation decreased. This same

Table V

Material SAE	Material AMS	Heating Temp. Degrees Fahr.	T. S. p.s.i.	Yield Point p.s.i.	Elong. Per Cent in 2 Inches	Red. Area Per Cent	Brinell	Charpy Impact Ft.-Lbs.
3312	6250	2350	142,000	108,000	18	65.3	311	32
3312	6250	2350	154,000	118,500	17	61.5	311	33
3312	6250	2350	137,200	105,500	17	64	302	34
Average			144,400	110,666	17.3	63.6	308	33
3312	6250	2500	143,000	109,500	14	47	311	32
3312	6250	2500	144,100	113,500	15	48.5	311	33
3312	6250	2500	159,000	126,500	12	50.5	302	34
Average			148,700	116,500	13.6	48.6	308	33

trend is also true in the steels S.A.E. 4140, S.A.E. 4340, and S.A.E. 3312, as shown in Tables III through V.

Figs. 7 through 10 were made of the tensile fractures and the facets are clearly seen in those bars which were heated to 2500 degrees Fahr. (1370 degrees Cent.). All tensile fractures heated to 2350 degrees Fahr. (1290 degrees Cent.) show a star-shaped fracture with the exception of S.A.E. 3312, which has a cup and cone effect. All fractures at 2500 degrees Fahr. (1370 degrees Cent.) are of the cup and cone fracture. The facets shown in all the photographs are not as pronounced as is seen in the actual fractures, due to the difficulty of photographing fractures. To make them more distinct, we had them magnified two diameters for the photographs.

This lowering of the per cent elongation and reduction of area by high temperature exposure is in agreement with the publication of one of the authors.[1] In this publication, tensile tests made on S.A.E. 3145 steel showed a lowering of the per cent elongation and reduction of area with exposure to temperatures of 2450 degrees Fahr. (1345 degrees Cent.) and above. In this same work no similar effect from overheating of plain carbon steel was found.

In addition to indicating some impairment of physical properties, the appearance of facets in the fracture suggests that the forging was done at very high temperatures.

DISCUSSION

Written Discussion: By Charles F. Sawyer, research engineer, Carnegie Institute of Technology, Pittsburgh.

The authors state that facets on the fractured surface of alloy steel forgings are not likely to be encountered in plain carbon steels, or in alloy steels

[1]W. E. Jominy, "Burning and Overheating of Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 16, 1929, p. 372.

in the normalized condition. The writer would like to present an experience with plain carbon steel press forgings.

About 3 years ago facets were observed in the fracture of longitudinal tensile test specimens from large press forgings. The forgings were made from S.A.E. 1035 steel which was received as 16-inch square blooms and subsequently press forged to 16-inch diameter rounds. After forging, the bars were air-cooled and then normalized.

Longitudinal tensile tests were made and the ductility values were lower than normally expected. The fractured surfaces of these test specimens showed facets which appeared as described by the authors. A review of the practice used seemed to indicate that (a) the forging temperature was too high, and (b) the reduction by forging was not large enough.

In an effort to eliminate these facets and to increase the ductility, the bars were renormalized. After this treatment the tensile test fractures showed no facets and the ductility values were quite normal.

A double normalizing treatment was then incorporated in the practice and the forging temperature was kept as low as possible. No more trouble was experienced with either low ductility or facets in the fractures of the tensile tests.

Written Discussion: By A. F. Christian, chief metallurgist, Wyman-Gordon Co., Harvey, Ill.

This article has been studied with interest, since the information contained therein confirms in a large degree that obtained by a considerable amount of research and an enormous amount of production line fracture testing with which the writer was directly concerned. There are, however, several points with which the writer is not in complete agreement.

The fracture test was developed primarily, not as an indicator of questionable quality associated with high forging temperatures, but in order to conserve large amounts of alloy steel at a time when such material was extremely critical. The physical tests in use could properly pick out overheated or unsatisfactory forgings but the material required for such tests amounted in some cases to 20 pounds per forging, whereas the fracture test required less than a fifth of this amount. At least half a million such tests were conducted by the forge shop with which the writer is associated and hundreds of tons of alloy steel were saved by the substitution of fracture testing for the regular tension and impact tests.

The presence of a few facets on a fracture sample does not mean that the quality of a forging is questionable. Just as specifications show minimum values for physical properties, so there are maximum amounts of grain coarsening or facet formation beyond which material should be scrapped. With the exception of tensile strength, which is not influenced by grain coarsening, the average physical property values shown in Table IV for S.A.E. 4340 steel heated to 2500 degrees Fahr. are above the average specification set up by several aircraft engine manufacturers. A comparison is as follows:

	Tensile Strength	Yield Point	Elongation Per Cent	Reduction in Area Per Cent	BHN
Average Spec.	140,000	120,000	15	50	302-341
Table IV	139,475	126,875	19.2	56.3	302

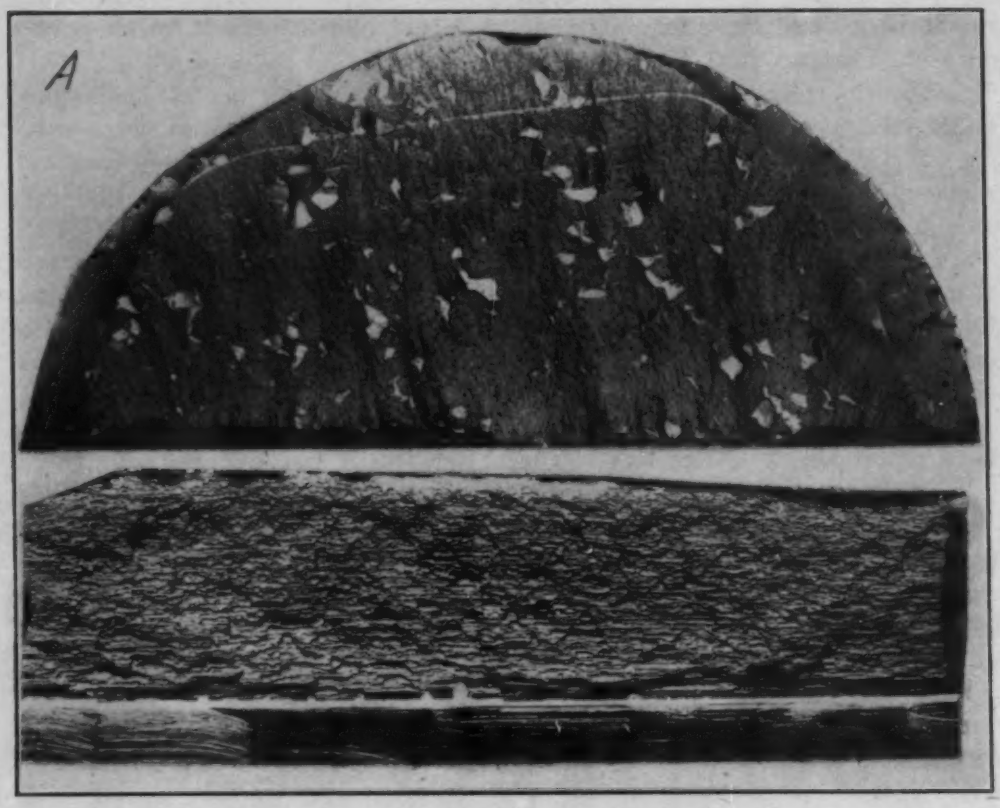

Fig. A—Comparison of Transverse and Longitudinal Fractures. Acceptable.

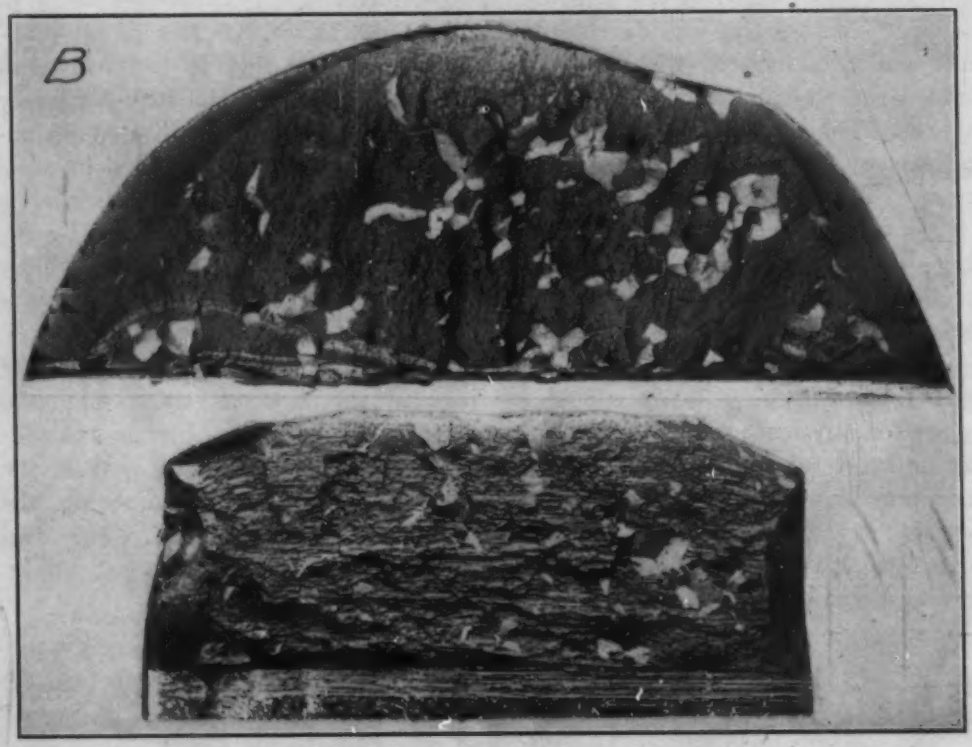

Fig. B—Comparison of Transverse and Longitudinal Fractures. Not acceptable.

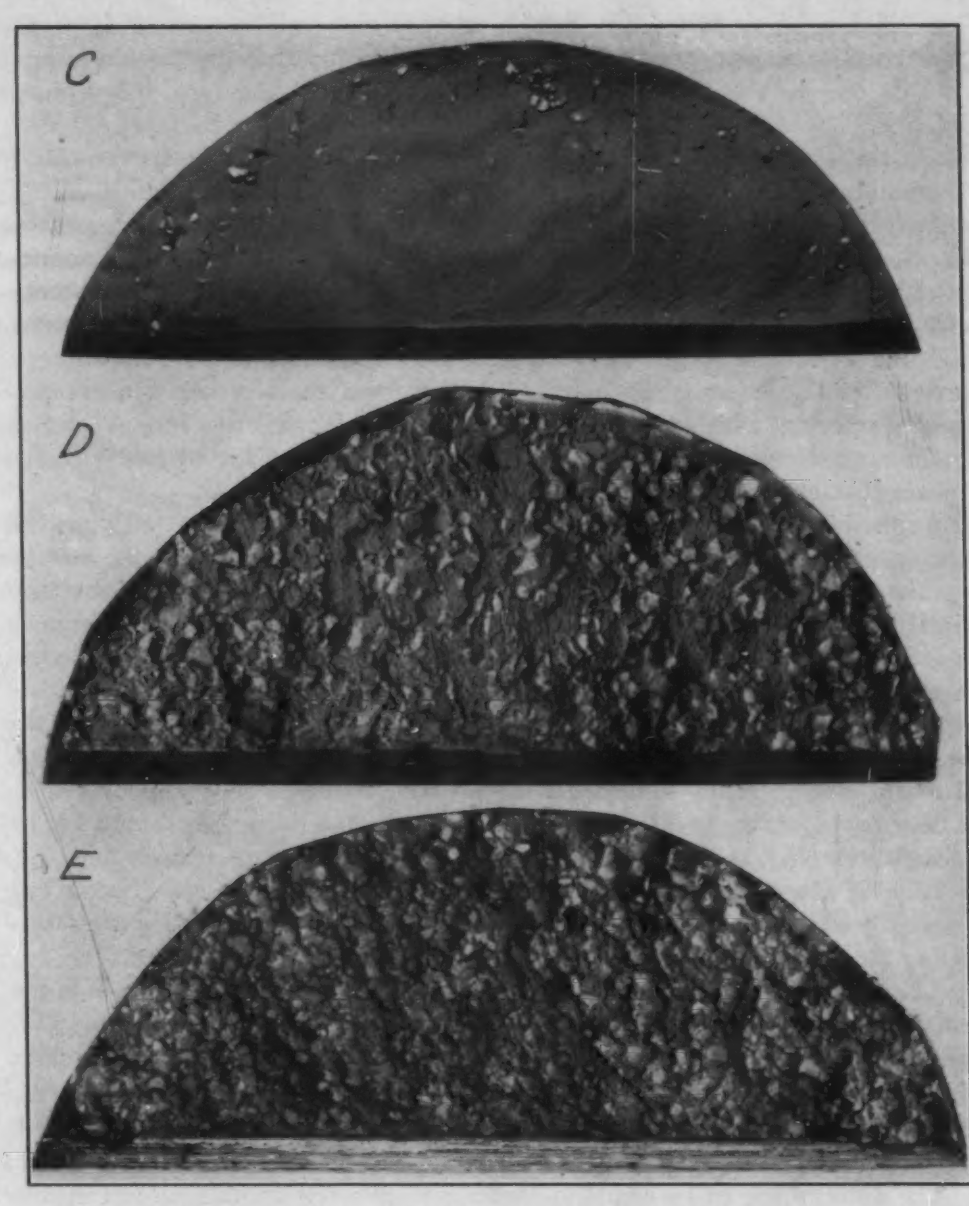

Fig. C—Effect of Hardness on Quenched and Drawn Fractures. Oil-quenched, not drawn, 495 Brinell.

Fig. D—Same as Fig. C. Drawn to 415 Brinell.

Fig. E—Same as Fig. C. Drawn to 341 Brinell.

This does not mean that we would allow a maximum forging temperature of 2500 degrees Fahr. for we have found, contrary to the statement in the paper, that time at heat does have an effect on the amount of grain coarsening and the attendant physical properties. We have insisted on a maximum furnace temperature of 2325 degrees Fahr. and a stock temperature of 2250 degrees Fahr. Even at these temperatures we have found grain coarsening when the stock was allowed to remain in the forging furnace during hammer repairs.

This opinion is supported by British metallurgists who commented on the appearance of facets on fractures as follows: "We feel that the trouble is most probably due to the fact that the shafts were soaked too long before forging . . . ."

At the top of page 544 the authors state that "They (facets) appear only if the fracture is transverse to the grain flow". Our first tests were made longitudinally and the original fracture specification, as set up by one of the major aircraft engine producers and approved by the Navy Bureau of Aeronautics, stated that material would not be acceptable if facets were evident on longitudinal fractures. Although this was a definite point of distinction between good and rejectable forgings, the specification was later based on transverse fractures, as this latter type was more sensitive and revealed minor degrees of grain coarsening which were allowable but not desirable. Figs. A and B show a comparison of transverse and longitudinal fractures for varying amounts of grain coarsening.

In the last paragraph on page 544 the authors state that ". . . . all quenched structures may be used to reveal facets . . . ." We have found that the amount of grain coarsening shown is dependent on Brinell hardness and that above 430 Brinell the background changes from a fibrous to a silky type which disguises the facets almost entirely. Figs. C, D and E show a severely coarsened sample as quenched, drawn to 415 Brinell, and drawn to 341 Brinell. The as-quenched sample shows only slight evidence of grain coarsening, while the fracture at 341 Brinell shows considerable more coarsening than at 415 Brinell. 302 to 341 Brinell seems to be the most sensitive hardness range.

We have found, as have the authors, that facets can be eliminated by additional forging at temperatures below the facet-forming temperature. A reduction of approximately 25 per cent will accomplish this result. Facets can be partially eliminated by simply reheating to a temperature just below the facet-forming temperature.

The facet-forming temperature of 2375 degrees Fahr. for S.A.E. 4340 is too high. We have produced facets experimentally at 2350 degrees Fahr. and evidence indicates that some heats will start to coarsen at just over 2300 degrees Fahr. Heats of almost identical analysis and grain size will react differently, so that one heat may show no facets while another will show considerable grain coarsening when heated exactly the same.

The authors are to be complimented for their work on this subject and it is hoped that further work will be done in an effort to determine the exact nature and mechanism of facet formation.

Written Discussion: By R. B. Schenck, Buick Motor Division, General Motors Corp., Flint, Mich.

This excellent paper is of more than usual interest to me in view of similar tests recently conducted at the Buick plant.

Our work was confined to A.M.S. 6415 (S.A.E. 4340) and consisted of a series of fracture, tensile and Izod tests on 3-inch rounds following various combinations of treatments which involved overheating for forging, overheating without forging, normalizing, annealing, quenching and tempering (Brinell 400). All samples were forged from 7 inch square to 3 inch round and fin-

ished at 1700 to 1800 degrees Fahr. Fracture tests were made transverse to the grain flow and tensile and Izod tests were taken at half-radius of the 3-inch round.

Following is a summary of our results:

1. After forging from 7 inch square to 3 inch round at 2300 degrees Fahr. and immediately reheating (without cooling below the finishing temperature) by 50-degree steps to temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling, normalizing, annealing, quenching and tempering, facets appeared in the fracture test on the sample heated to 2350 degrees Fahr. and increased progressively in size and number with increase in temperature. This progressive increase in facets was accompanied by a downward trend in elongation, reduction of area and Izod, with a marked drop in each case between 2500 and 2550 degrees Fahr. However, this decrease in mechanical properties, even at 2550 degrees Fahr., seems hardly sufficient to seriously impair the quality of the material.

2. After forging from 7 inch square to 3 inch round in 50-degree steps at temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling, normalizing, annealing, quenching and tempering, no facets were found in the fracture tests. All fractures were identical in appearance, and there was no significant difference in elongation, reduction of area and Izod values.

3. Fracture tests after forging at 2300 degrees Fahr. and immediately reheating to temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling with no subsequent treatments showed a progressive increase in grain size with no facets.

4. After forging at temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling with no further treatment, the fractures were identical in appearance and showed no facets.

5. Forging at 2300 degrees Fahr. and immediately reheating to temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling, normalizing and annealing resulted in fractures identical in appearance with no facets.

6. Forging at temperatures of 2200 to 2550 degrees Fahr. followed by slow cooling, normalizing and annealing resulted in fractures identical in appearance with no facets.

In addition to the tensile and Izod tests we have a series of fatigue tests underway which have not been completed. It is hoped that the fatigue results will show something conclusive.

Referring to Tables II to V in the paper under discussion, it is interesting to note that, while all four steels showed lower elongation and reduction of area when facets were present, only the 4340 showed lower Charpy values. Also of interest is the resistance of Nitralloy G to facet formation.

I would like to ask the authors whether they have any data on the effect of prolonged heating on formation of facets, such as would occur due to a delay in removing the stock from the forge furnace.

I would also like to inquire whether any information is available regarding the uniformity of the temperature of facet formation from heat to heat of the same composition. Since the grain coarsening temperatures vary considerably with the same chemistry, it would seem that facets might follow a similar pattern.

It is indeed gratifying to find such close agreement in the results of two independent investigations.

Written Discussion: By E. O. Dixon, chief metallurgical and mechanical engineer, Ladish Drop Forge Co., Cudahy, Wis.

The authors are to be complimented for a timely paper on a subject about which new information is sure to be of great benefit to all. Any additional light which can be cast on a new method of improving quality or preventing material loss through more effective detection or elimination of defects is, of course, a contribution to progress.

The authors' principal premise, i.e., that excessive occurrence of coarse facets or polyhedral grains caused by exposure of steel, apparently of aircraft grades, to forging temperatures of 2400 to 2500 degrees Fahr. is accompanied by lowering of ductility, will be readily accepted by many metallurgists who have studied this phenomenon lately. Fortunately, these temperatures are distinctly in excess of the levels commonly encountered in best forging practices.

It is noted that with but one exception tensile test data are given for only two temperatures, i.e., the extremes used in the investigation. It would be interesting to know whether the authors ascertained in a similar manner the effect on ductility to be associated with the occurrence of smaller proportions of facets or polyhedral grains. Haworth and Christian, in their excellent paper,[2] reported no reduction in ductility until a considerable concentration of facets had developed, which corresponds with this discusser's experience on many mill heats of aircraft grades of S.A.E. 3312 and S.A.E. 4340. It is also now reported in this connection that there is a wide range of sensitivity of various mill heats toward formation of facets or polyhedral grains with respect to temperature. In an extended study of this point a total of over 100 mill heats was subjected each to four different high temperatures. It was found that temperatures necessary to show incipient polyhedral grain formation covered a range of over 150 degrees Fahr. It was evident, furthermore, that these temperatures are to some extent determined by steelmaking conditions, because among six steelmaking sources involved, there were found distinctly different minimum temperatures of polyhedral grain formation. Table A summarizes these data, which were obtained by fracturing samples cut from billets as received from the mill, which were subjected to various high temperatures, then normalized, quenched, and tempered.

It may also be pointed out that polyhedral grains, by their peculiarly regular geometric shape, seem necessarily to have formed after all severe plastic deformation has occurred. It does not seem possible that the planar and rectilinear characteristics indicated by their appearance could persist through the three-dimensional plastic disturbances which are inherent in the drop forging process. Consequently, it must be assumed that some form of recrystallization takes place after metal flow has ceased, which in many cases is, without question, at a relatively low temperature. In at least one instance, polyhedral grains were found in a forging which was made in two hammers with one heating of the billet. In this case, plastic flow occurred at the site of the polyhedral grains

[2]R. D. Haworth, Jr. and A. F. Christian, "Fracture Testing of Alloy Steels for Aircraft Engine Forgings", *Proceedings*, American Society for Testing Materials, Vol. 45, 1945.

at a temperature known to be well under 2200 degrees Fahr.

Because of such observations and considerations there arises question as to whether these facets or polyhedral grains are not manifestations of a phenomenon of far more extensive interest than mere bright spots which we hope we shall not see when we make fracture tests on aircraft forgings.

Table A
Number of Mill Heats From Various Sources of A.M.S. 6250 Steel Showing Incipient Polyhedral Grain Formation at Various Temperatures

Source	2250 Degrees Fahr.	2275 Degrees Fahr.	2300 Degrees Fahr.	2325 Degrees Fahr.	None at 2325 Degrees Fahr.	Total
A	1	1	3	2	6	13
B	2	4	3	2	7	18
C	0	1	0	0	0	1
D	7	23	4	3	11	48
E	0	5	3	0	2	10
F	0	1	0	0	0	1
Total	10	35	13	7	26	91

Written Discussion: By John Welchner and J. J. Russ, Metallurgical Department, The Timken Roller Bearing Co., Steel and Tube Division, Canton, Ohio.

This paper has been read with extreme interest and the authors are to be complimented on adding specific data to the rather meager knowledge of fracture testing.

If the mere appearance of facets in the fracture of an aircraft forging is to be used as a criterion of acceptability of such forgings, certainly considerable more time and energy should be expended to gather additional quantitative data concerning this problem. Not only is it desirable to know more about what factors affect fracture appearance and how these factors can be controlled, but it is imperative to determine if a facet-containing fracture is really detrimental to the performance of an aircraft engine member. Admittedly, cases have been cited where extremely coarse facets in the fracture (where evidence of actual "burning" of portions of the steel might easily have been removed in machining) were thought to be the cause or a contributing cause for failure of an engine part. However, it is doubtful whether the present paper, or other investigational work on this subject, has conclusively shown that a facet-containing fracture is detrimental to the performance of the part, or indeed, if it even impairs any physical property of the steel when occurring in amounts less than a limiting value.

In working on this problem in our laboratory in the past several years, we have developed certain experimental data which may be of interest to others. While our investigations covered quite a few different heats of steel, only the A.M.S. 6250 (S.A.E. 3312) and A.M.S. 6415 (S.A.E. 4340) types were included. A summary of some of these results is offered for the record.

Since the overheated forging problem was originally most prevalent with A.M.S. 6250-type forgings, studies to determine the maximum permissible heat-

Table B

Heated at Temperature Indicated for 2 Hours, Followed by 1725 Degree Fahr. Normalize, 1175 Degree Fahr. Temper, 1700 Degree Fahr. Pseudo-Carburize, 1475 Degree Fahr. Oil Quench, and 350 Degree Fahr. Temper

Temperature Degrees Fahr.	Charpy Ft.Lbs.	Charpy BHN	Charpy Per Cent C	Charpy Fracture	Izod Ft.Lbs.	Izod BHN	Izod Fracture	1½-Inch Square Fracture
				Heat A				
1900	42	302	0.11	OK	64-62-64	321	OK	OK
	40	302		OK	66-65-66	302	OK	
2100	39	302	0.11	OK	64-66-68	321	OK	OK
	39	311		OK	64-66-67	321	OK	
2200	39	302	0.10	OK	62-64-64	321	OK	OK
	35	302		OK	64-67-66	311	OK	
2300	40	302	0.11	OK	68-69-68	302	*BL	BL
	39	302		OK	68-68-68	302	BL	
2400	43	302	0.11	Facets	68-70-68	302	Facets	Facets
	..	...		Facets	66-70-68	302	Facets	
				Heat B				
1900	28	363	0.13	OK	44-45-44	363	OK	OK
	28	363		OK	45-45-45	363	OK	
2100	40	302	0.10	OK	73-75-75	321	OK	OK
	42	302		OK	75-76-76	302	OK	
2200	34	321	0.11	OK	50-52-50	321	OK	OK
	33	321		OK	55-52-51	321	OK	
2300	42	352	0.12	OK	55-55-55	321	OK	OK
	40	341		OK	60-64-64	311	OK	
2400	34	302	0.12	Facets	62-62-62	302	Facets	Facets
	34	302		Facets	60-62-62	302	Facets	

*BL—Indicates Borderline

ing temperature to prevent a facet fracture were concentrated on this analysis steel. In one such investigation, two production, electric furnace, aircraft heats of the following chemical analyses were chosen for the tests:

Heat	C	Mn	P	S	Si	Cr	Ni
A	0.11	0.38	0.015	0.017	0.29	1.33	3.59
B	0.10 to 0.13	0.39	0.014	0.024	0.28	1.32	3.60

Duplicate 1½-inch square by 6-inch long samples were obtained from 6 by 6-inch billet stock. One set of samples was heated to each of the following temperatures: 1900, 2100, 2200, 2300, and 2400 degrees Fahr. Heating was done in a Globar-type, platinum-rhodium couple, L & N controlled furnace. The specimens were maintained 2 hours at temperature followed by an air cool. All pieces were then subjected to a 1725 degree Fahr. normalize, 1175 degree Fahr. temper for 3 hours, 1700 degree Fahr. pseudo-carburize for 12 hours and pot cool, 1475 degree Fahr. reheat for oil quenching, and a 350 degree Fahr. temper. This duplicates the heat treatment given most aircraft gears of this analysis. Since some users of this type aircraft steel claimed an impairment of impact value with the occurrence of facets in the fracture, duplicate 3-notch Izod and duplicate Charpy specimens in addition to fracture ratings of these tests and fracture examination of the 1½-inch square samples were obtained. The 1½-inch squares were abrasive-saw notched and fractured in a tensile machine. These results are given in Table B and shown graphically in Fig. F.

From these tests it is apparent that:

(a) High heating or processing temperatures did not impair the impact strength of this steel when tested in the final heat treated condition even when facets appeared in the fracture. This is in agreement with the authors' findings.

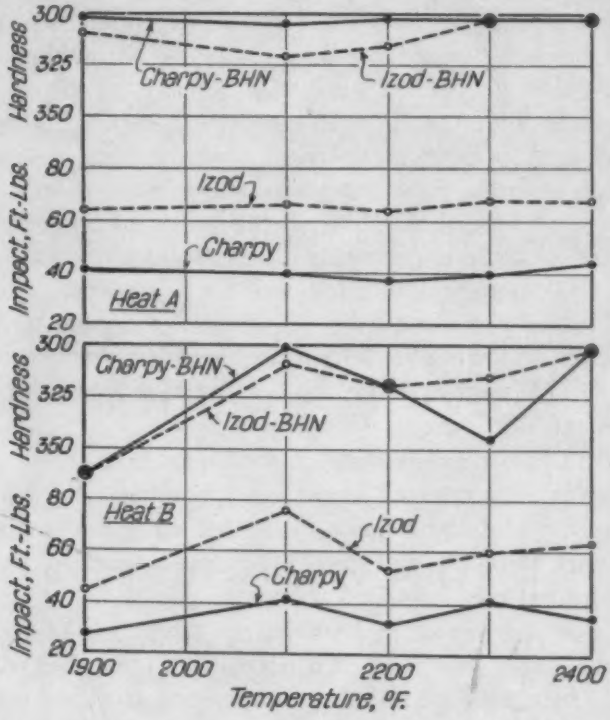

Fig. F—Heated at Temperature Indicated for 2 Hours Followed by 1725 Degree Fahr. Normalize, 1175 Degree Fahr. Temper, 1700 Degree Fahr. Pseudo-Carburize, 1475 Degree Fahr. Oil Quench and 350 Degree Fahr. Temper.

(b) The impact values were affected by small variations in carbon content and Brinell hardness readings.

(c) Both heats when heated to 2400 degrees Fahr. for 2 hours produced facets in the fracture, indicating that the facet-forming temperature was somewhere between 2300 and 2400 degrees Fahr. This temperature is somewhat lower than that found in the paper under discussion.

It is pertinent to note that the foregoing data were obtained on material which was only heated to forging temperatures but actually received no hot work; consequently, the interpretation of results must be subject to this limitation.

This same type steel was further studied in several investigations, when suitable samples from 36 production heats of electric furnace, aircraft grade, were heated ¾ to 1 hour at 2300 degrees Fahr. followed by the gear heat treatment previously described. Because our interest at the time these experiments were started was not focused on fracture testing alone, only the single heating

time and temperature was employed for most heats. No attempt was made to determine the facet-forming temperature in these tests. These results substantiated the authors' and our previous findings that each heat gave a facet-free fracture after heating at 2300 degrees Fahr. The Izod values of 16 of the heats so tested, however, ranged from 41 to 76 foot-pounds. This variation was, within the experimental error, inversely proportional to the Brinell hardness of the test piece in all but one instance. In the exception, the Izod value was higher than the hardness would predicate and was left without a suitable explanation.

With respect to the A.M.S. 6415-type steel, 28 electric furnace, aircraft heats were examined in several different investigations. Again, no specific attempt was made to locate the exact facet-forming temperature for these heats, but enough data were developed to bracket this value for production heating purposes. In the first of these experiments, billet samples of 15 heats were heated for ¾ hour at 2300 degrees Fahr. followed by a sil-o-cel cool to room temperature. Duplicate 2-inch square by 4½-inch long specimens from each heat were then given the following heat treatment prior to fracture testing:

1. Heated at 1700 degrees Fahr. for 3 hours and air-cooled.
2. Tempered at 1200 degrees Fahr. for 3 hours.
3. Reheated to 1525 degrees Fahr. and oil-quenched.
4. Tempered at 1050 degrees Fahr. for 3 hours to 302/341 BHN.

Fractured surfaces of the 3-notch Izod specimens as well as those of one of the 2-inch square samples from each heat were examined for facets. None of the heats yielded entirely facet-free fractures.

In another test, samples of 13 heats of the same type steel were heated at 2250 degrees Fahr. for 1 hour and then normalized, quenched and tempered to 321-363 BHN. These produced all satisfactory or facet-free fractures. These findings coupled with those previously obtained on the 15 heats indicate that the facet-forming temperature for this analysis lies between 2250 and 2300 degrees Fahr. This seems to be somewhat at variance with that of the authors' permissible value of 2375 degrees Fahr. for this analysis steel.

Numerous laboratory and shop tests with which we have been connected on several occasions suggest that certain other factors affect the ultimate condition of the heat treated fracture when the facet-forming temperature has been exceeded. It appears that forgings made from steel which has been heated below a certain maximum temperature will in all instances produce a facet-free fracture. The converse is not necessarily true. Steel which has been heated in the facet-forming range either will or will not produce a facet-containing fracture depending upon one or more of the following conditions:

(a) Finish hot forming temperature—if the final recrystallization of the metal, which constantly accompanies hot working, occurs below the limiting temperature for the type steel involved, a completely facet-free fracture will be produced. This has been emphasized by the present authors.

(b) The amount and type of hot work depending upon the design of the part—where certain sections of forgings receive a nil amount of hot deformation, only little benefit of recrystallization will be gained. This

is illustrated by several aircraft gear forgings in which quite often the fracture of the much worked teeth section is satisfactory while that of the minimum deformed shaft portion contains facets. This condition exists even though all sections of the original bar stock were heated to the same forging temperature.

(c) Time at temperature—instances have been reported where a few minutes at high temperature produced facet-free fractures whereas several hours at the same temperature did not.

(d) Rate of forging—as pointed out by the authors, the temperature of the steel can either be increased or decreased during the hot forming operation, depending upon size of forging, relation of mass to cooling surface, etc.

(e) Rate of cooling from the forging temperature—a rapid cool tends to eliminate facets in the fracture while a slower cool from the identical forging conditions will in some instances yield a facet fracture.

Written Discussion: By G. R. Barrow, chief metallurgist, Canton Drop Forging & Mfg. Co., Canton, Ohio.

It is unfortunate that the authors were not able to complete the work and publish their paper several years ago. The information which they obtained would have been of great benefit to the forging industry. Throughout any discussion on overheating, it is well to consider the magnification at which the sample is viewed and the degree of coarseness observed. The writer would like to make a few comments on parts of this paper.

Reference is made to the work done by the authors on the effect of extensive soaking time relative to the size and number of facets. We are in agreement with their findings if it is carried out above the facet-forming temperature. We found, particularly on the A.M.S. 6250 and 6412 analyses, that any abnormal soak, reasonably below the facet-forming temperature, contributes to the formation of facets and the general coarseness of the steel. In other words, time at temperature is very important at the normal forging temperature.

There is general agreement on the top forging temperatures as listed by the authors. Work carried out at our plant on the A.M.S. 6250 and 6412 steels pointed to the fact that aside from the facets, the first indication of overheating showed up on the Izod impact test. A lowering of the elongation and reduction of area was noted but the Izod test seemed to be a better barometer to use.

If we were to list the factors in the order of their importance, which contribute the most to overheating and the formation of facets, they would be: 1. Maximum work temperature, 2. Amount of work done on the piece, 3. Time at temperature, 4. Number of reheats, 5. Design of the forging, 6. Finishing temperature, 7. Inherent heat characteristics of the steel and other miscellaneous factors.

Written Discussion: By Michael G. Corson, consulting metallurgist, 603 West 111th St., New York.

About 3 years ago I encountered the necessity for examining a continuous flow of forgings called "spiders"—the most important single part of the Hamilton variable pitch propellers. The Hamilton Co. itself established a rather efficient practice of making those spiders—with the assistance of the

Wyman-Gordon Co. which supplied the forgings. The latter were made in a series of nine forging operations, and, as I understand it, there hardly were any cases of rejection of the castings either in the laboratory or by the machine shop.

For a short time our company was getting spider forgings from the same source and none of these gave troubles in machining although the hardness was rarely lower than 415 Brinell. The steel was S.A.E. 4340.

Then we began to receive forgings from a different source. These were made in three forging operations. And for a time at least we had nothing but troubles. I decided then to run a series of full scale static load tests. The machined spiders were loaded statically with a pressure that could run to 600,000 pounds total. And I found that the spiders forged in nine operations stood a far heavier load and did not break—simply bent beyond the possibility of further loading. The forgings made in three operations stood a much lower elastic load and bent beyond the possibility of further loading at a lower load value, while those that actually possessed a higher hardness (415 to 427) broke and the fractures always exhibited clusters of facets. Each facet was about 0.05 inch in diameter.

In addition to facets the same spiders that broke in the tests exhibited rather sharply defined areas of striation. The average hardness in these areas was by no means different from that of the metal around them, but they caused incessant troubles in machining.

I do not know at what temperatures the steel ingots used in the production of the described forgings were forged into billets, but the temperature of the forging into spider stocks was never higher than 2300 degrees Fahr. Consequently the presence of facets and the "hard" areas that accompanied them could not be traced to high temperature. And I believe that the true cause of their formation lies in the excessive speed of shaping. If the authors of the paper under discussion had to go to very high temperatures to produce facets, it might be due to the very modest shaping speed with which experimental forging for laboratory tests is being done.

I may add that the "fingers" broken off of the spiders in my static tests had always a lower density in comparison to such fingers as were cut off for the purpose of making test bars. I would assume, as a matter of hypothesis, that the excessive speed of forging caused by the facets, by simply separating a number of grains from their neighbors along some more or less sharply defined planes.

Oral Discussion

H. E. Moser:[3] We found facets present in 4130 forgings after heat treatment. Facets were due to overheating before forging. The questions I wish to ask are, one, whether the facets can be identified with the microscope at over 100 magnification? The other question is, when the physical properties have been lowered by the presence of facets, is there any satisfactory method of reclaiming this material?

[3]Chief metallurgist, McCord Corp., Detroit.

We found that the presence of facets also reduced the impact properties.

G. B. TRUMBLE:[4] This work done by Messrs. Strohm and Jominy is of very great interest to all drop forgers. Of particular interest is the work done in conjunction with the S.A.E. 3312-type steels (A.S.M. 6250 and 6254), as well as the S.A.E. 4340 steels. There have been, over the last few years, quite a few discussions concerning the appearance of facets in fracture testing for overheating. Those temperatures given indicating temperatures at which these facets form are rather startling as compared with some of those that have been observed by other drop forgers. For instance, if we take just a minute or two on this S.A.E. 3312-type steel, I think that there has been considerable work done, and am merely offering this for future comment, as to temperatures not being the whole story in regard to the formation of facets, it being the consensus of opinion from information gathered by myself that there are three factors involved—namely, time, temperature and amount of work done. We have seen facets form with forging furnace temperatures controlled by Leeds and Northrup Ray-O-Tube equipment at 2200 degrees Fahr., with actual forging multiple checked by optical pyrometry not over 2200 degrees Fahr. This S.A.E. 3312-type steel in particular has shown a variance and peculiar behavior in regard to facets in fracture tests that has been very wide and varied and we have one of the steel mill representatives here who has done some work along this line. We have seen these facets form at forging temperatures beneath those indicated by the paper. This has also been observed on S.A.E. 4340 steels, particularly in the portions of a forging where there has been little or no work done, so that again the three factors of time, temperature and amount of work seem to be an indicator as to the occurrence of this particular condition.

I would be interested in comments of other people, particularly in reference to the S.A.E. 3300 series, because of the controversy over this particular grade of steel on gears used in aircraft engines. There are some indications that there is a difference of the susceptibility for this particular condition of facet forming between different heats of steel. This has been accepted by some steel companies and thoroughly denied by others.

F. E. HARRIS:[5] I have enjoyed this very interesting discussion on forging temperatures.

We ran similar tests on 4340, and found 2300 degrees Fahr. to be the highest safe temperature. At that temperature, regardless of the time held, we found no indication of facets when the heating was followed by slow cooling without forging.

Authors' Reply

We are pleased to receive the discussion provoked by this work. The case of facets observed by Mr. Sawyer in which they are seen in the normalized condition and eliminated by a second normalizing is new to us. The facets we observed remained after repeated normalizing and heat treatments, but could be eliminated by forging at lower temperatures.

Mr. Christian has done a great deal of work on the study of facets and we

[4]Formerly chief metallurgist, Atlas Drop Forge Co., Lansing, Mich. Now plant metallurgist, Tube Turns, Incorporated, Louisville, Ky.

[5]Furnace engineer, Buick Motor Co., Flint, Mich.

are glad to have his comments. He states that a quenched structure in the range of 302 to 341 Brinell is the most sensitive for tests for facets. This is the range in which we worked. To us the appearance of even a small number of facets indicates that the forging has been overheated. When we consider that it is the temperature of the piece after finish forging that causes facets, it is evident that it is necessary to start with a very high temperature unless no or very little mechanical work is done at the point of fracture. For some forgings, mechanical work is done so fast that heat is generated by the forging process itself and the temperature of the piece thus increased, but this usually does not heat the piece above its temperature as removed from the furnace, though occasionally it does.

Mr. Christian says he uses a furnace temperature of 2325 degrees Fahr. to obtain a stock temperature of 2250 degrees Fahr. This means that he is feeding steel through the furnace at a rate which does not permit the steel to come to the furnace temperature. When there is a work stoppage at the hammer and the steel is permitted to stay in the furnace longer than the prescribed heating time, the steel no doubt comes to furnace temperature. There may even be a surge in furnace temperature due to the fact that no cold steel is added to the furnace load under these conditions. Although there may be a shop rule that furnaces are to be shut off under these conditions, our observations are that the workmen all too frequently fail to do this. It is under these conditions that we believe facets are produced, and what appears to be due to increased time is actually due to increased temperature.

It is our opinion that tests should be made in small accurately controlled furnaces and there is danger of erroneous conclusions if large production units are used since temperature variation in these furnaces can be very great. Whenever our tests were made in such furnaces it was only with a thermocouple inside of the test piece.

We believe that within the chemical variations of one steel composition it is possible to have a difference of perhaps 50 degrees Fahr. in facet-forming temperature and so are pleased to find that Mr. Schenck's results come that close to ours. Also Mr. Christian's reported experimental results on S.A.E. 4340 show facets at 2350 degrees Fahr. In reply to Mr. Schenck's question regarding the uniformity of the temperature of facet formation we have found no large variation in temperature with different heats of the same composition. Other metallurgists whose opinion we respect have told us they find large differences in facet-forming temperatures. Whenever we have heard of this we have asked for samples of the steel since we should like to study such samples. To date we have received no such samples.

We are glad to have Mr. Dixon's discussion and in reply to his question as to the physical properties at temperatures between the extremes used in the investigation we ran physical properties on S.A.E. 4640 steel at 2400 degrees Fahr. and listed these data in Table II of the paper. These results indicate that there is a drop in ductility as the forging temperature is increased in the facet-forming range. This drop seems to become progressively greater as the temperature approaches the burning range.

We are glad to find Mr. Barrow in good agreement with us. Our observa-

tions on the effect of time at heat is that it is relatively unimportant in the range of 15 minutes to 5 hours.

Messrs. Welchner and Russ have done very good work in studying facets. They do not report large differences in facet-forming temperatures due to unexplainable differences in steel heats. Such differences are reported by Christian and Dixon and mentioned by Trumble. Welchner and Russ question whether the presence of facets is really detrimental to the performance of an aircraft engine member. Everyone seems to be in agreement that a large number of facets should be cause for rejection. It would seem advisable to make further tests on parts having only a few facets—particularly fatigue tests. But until the steel with such fractures is known to be entirely satisfactory we believe it it better to be cautious about its use.

Mr. Corson suggests shaping speed as the possible cause of facets. We were able to obtain facets without any shaping whatever merely by heating above certain high temperatures. It is possible that shaping speed increases the temperature of the part above the facet-forming temperature.

In reply to Mr. Moser's questions regarding finding facets metallographically, we understand the English metallurgists often use an etch described by Austin in Special Report No. 14, The Iron and Steel Institute, 1936, p. 189. Austin claims to have moderate success on nickel-chromium, nickel-chromium-vanadium and nickel-chromium-molybdenum steels. The etching solution which is used at room temperature is an aqueous solution containing 10 per cent nitric acid and 10 per cent sulphuric acid. The polished sample is dipped 3 times in this solution for 30 seconds at a time and wiped with cotton wool in running water between dips. After the final dip it is dried and examined at 75 to 100 diameters. We prefer the fracture method for positive identification of facets.

Replying to Moser's second question, when the physical properties have been lowered by the presence of facets the material may be reclaimed by forging to the extent of a reduction of at least 20 per cent of the area at a temperature below the facet-forming temperature.

There seems to be complete lack of agreement on the temperature necessary to form facets on S.A.E. 3312 steel. Trumble says he has obtained them at 2200 degrees Fahr., Dixon's table indicates he obtained a small percentage of heats which showed facets at 2250 degrees Fahr., Welchner and Russ obtained facets at 2400 degrees Fahr., and we obtained facets at 2450 degrees Fahr. Mr. Trumble's observations are based on production experience and although Mr. Dixon does not say so, it appears that his table was obtained from production experience. We believe that production temperature measurements are not in general sufficiently accurate for careful work. Messrs. Preece, Hartley, Mayer and Nutting[6] in a paper which was published just recently tested 55 steels for incipient overheating temperatures. Out of the 55 they found 2 steels which showed incipient overheating at 2282 degrees Fahr., 7 at 2327 degrees Fahr., and the remaining 46 steels between 2372 and 2552 degrees Fahr.

It seems we are justified in concluding that the presence of facets is a good indication of overheating in forging and that the forging temperature should be lowered for the particular forging operation in which they occur.

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[6]Advance copy, The Iron and Steel Institute, March, 1946.

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